

Phase Transformations - 3

Amorphization

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Outline :

1. Amorphous systems: thermodynamics, structural and
(some) electronic properties
 - volume F.E. considerations
 - rôle of interfaces
 - rôle of strain
2. Phase transformations under irradiation (*open systems*) :
non-equilibrium phase diagrams
3. Simple mechanisms for amorphization via irradiation
4. Ion beams: a microscope to study the amorphous state
reveals
 - kinetics vs. thermodynamics
 - amorphization mechanism is not unique
 - rôle of (i) strain, (ii) interfacial energy

"Thermodynamics" of amorphous systems

T_g = "glass temperature"



"slow" atomic
movements (\sim nm/s)

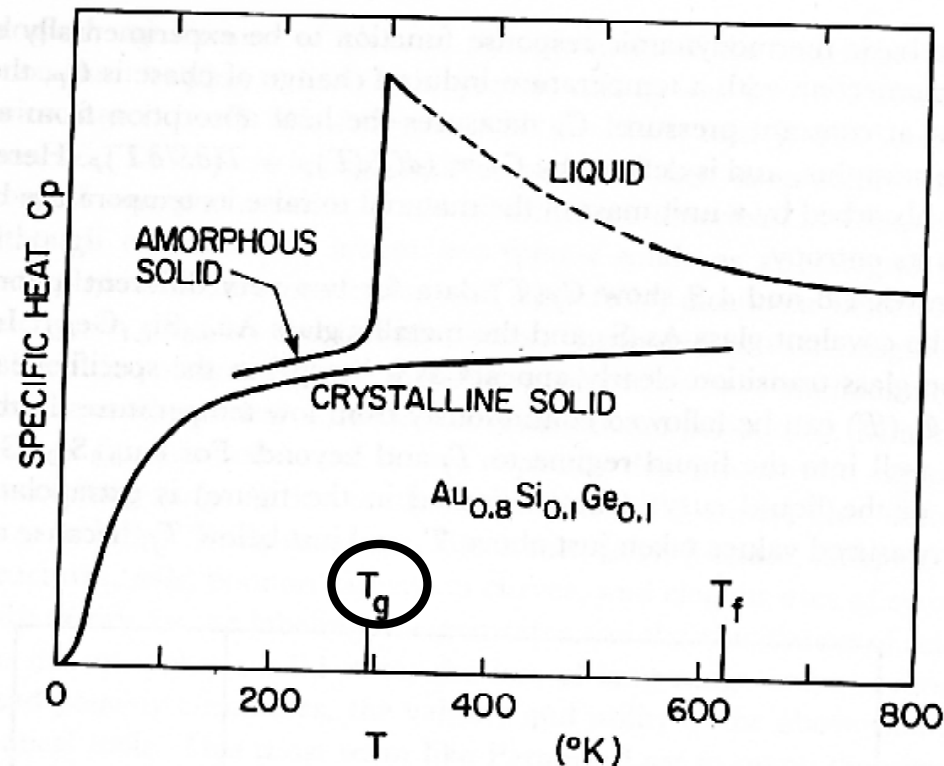
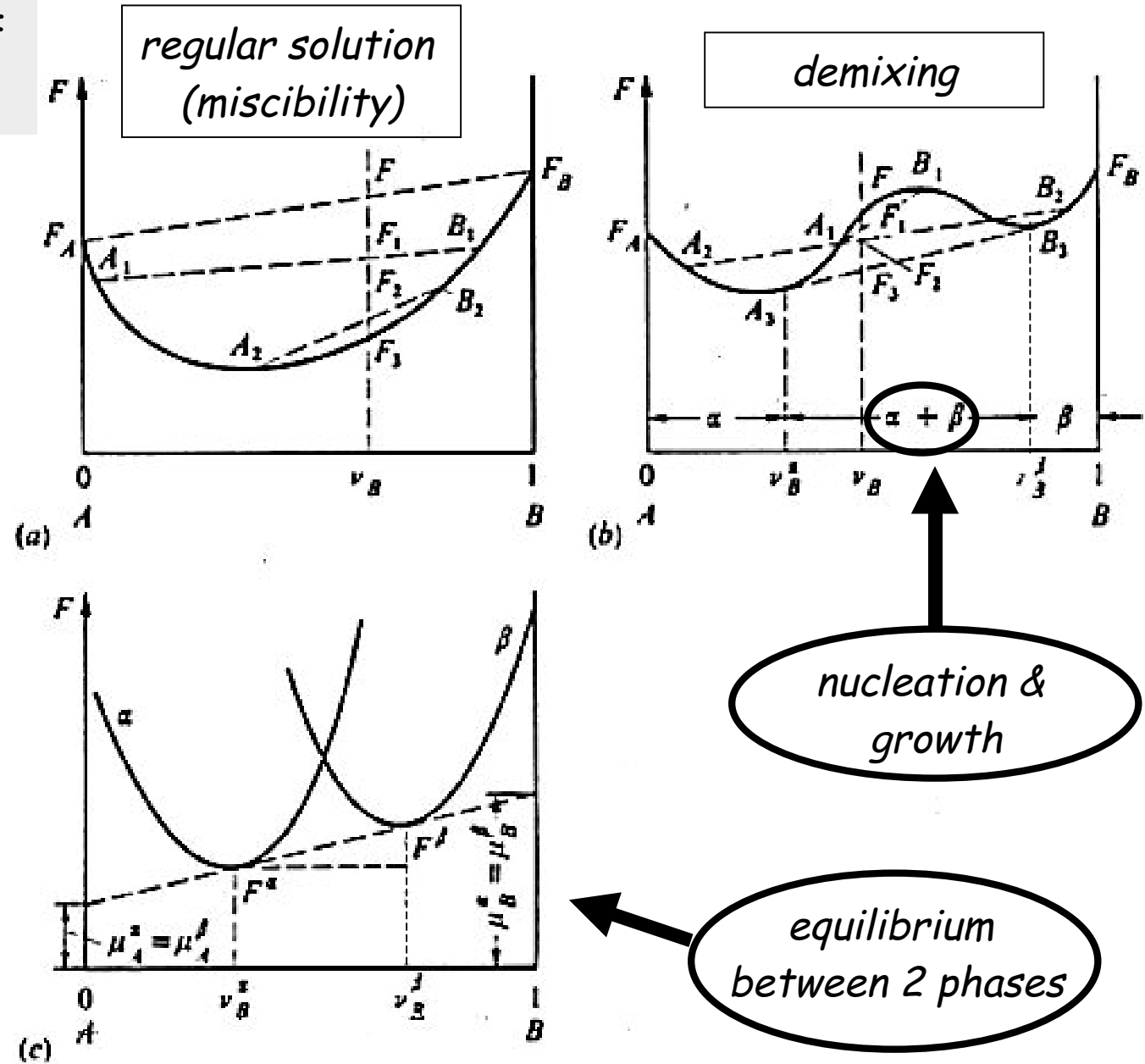


Figure 1.9 The specific-heat signature of the glass transition in a metallic glass (Chen and Turnbull, 1968).

A phase transformation is NOT necessarily a phase transition.
The glass transition is **not** a (thermodynamic) phase transition !
It is a *kinetic* transition (configurational freezing)

Thermodynamics of amorphous systems

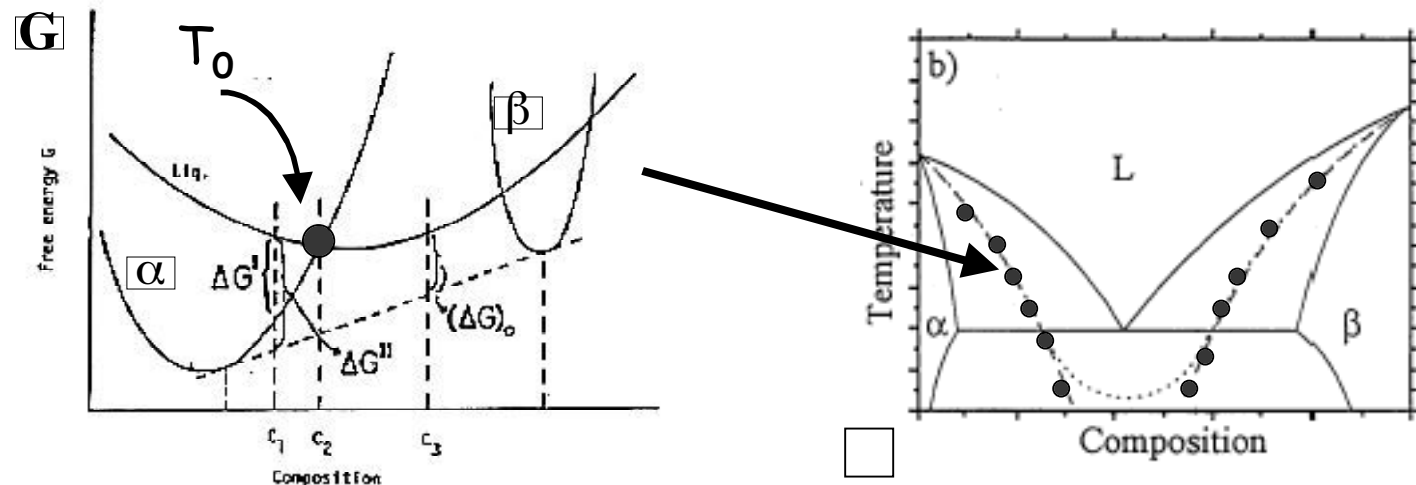
(a reminder on $\neq F.E.$ cases)



Thermodynamics of amorphous systems

Glass Forming Ability (GFA) near deep eutectic

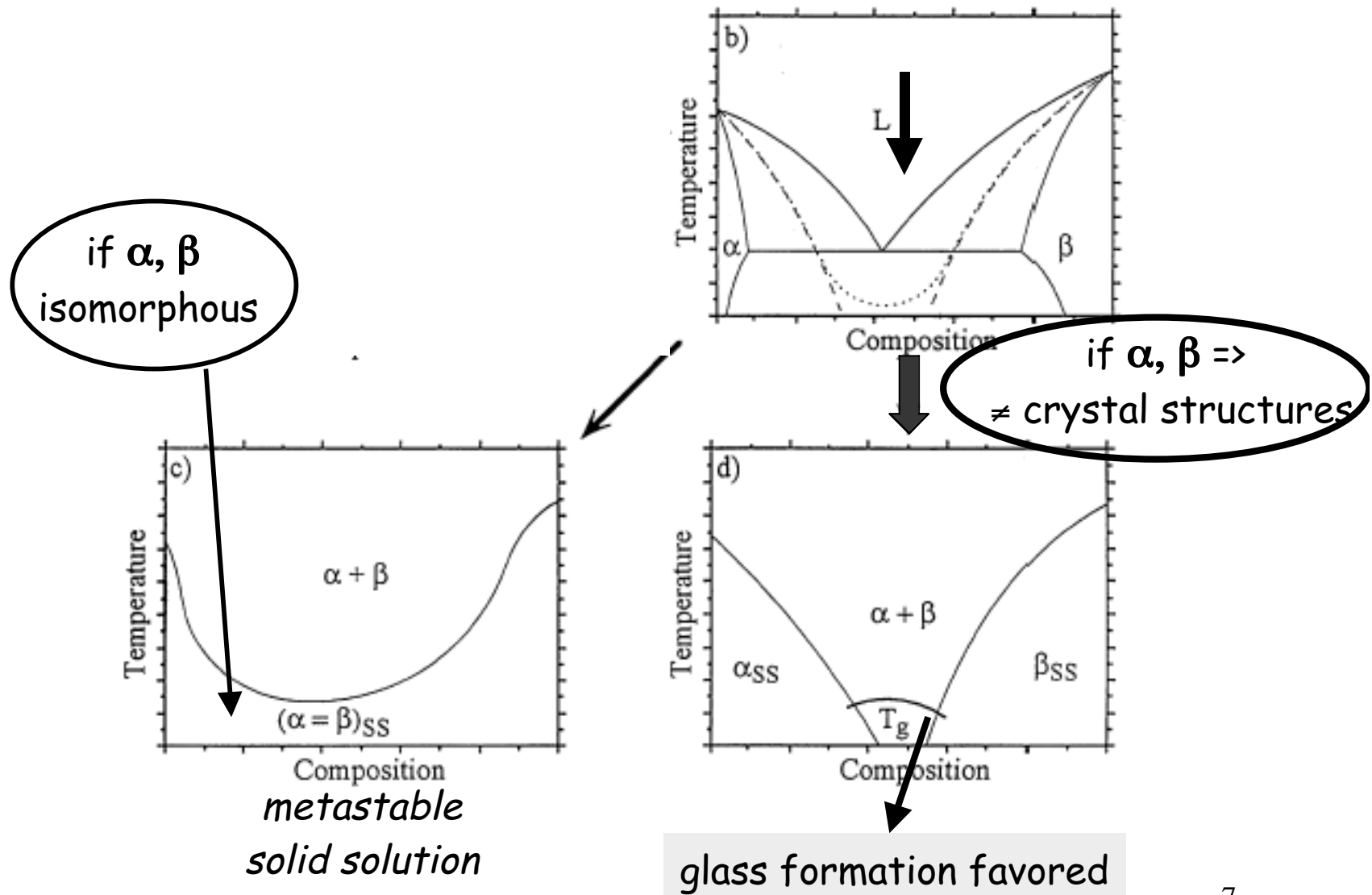
Define " T_0 " curve, such that
 $\Delta G(\text{liq}) = \Delta G(\alpha \text{ phase})$



Typical free energy conditions in the region of eutectic solidification at a temperature slightly higher than the glass transition temperature T_g

Thermodynamics of amorphous systems

Glass Forming Ability (GFA) near deep eutectic



Critical parameters for
Glass Forming Ability (GFA)

- Volume free energy
- Competition between \neq structures

Demixing condition
(nucleation & growth)

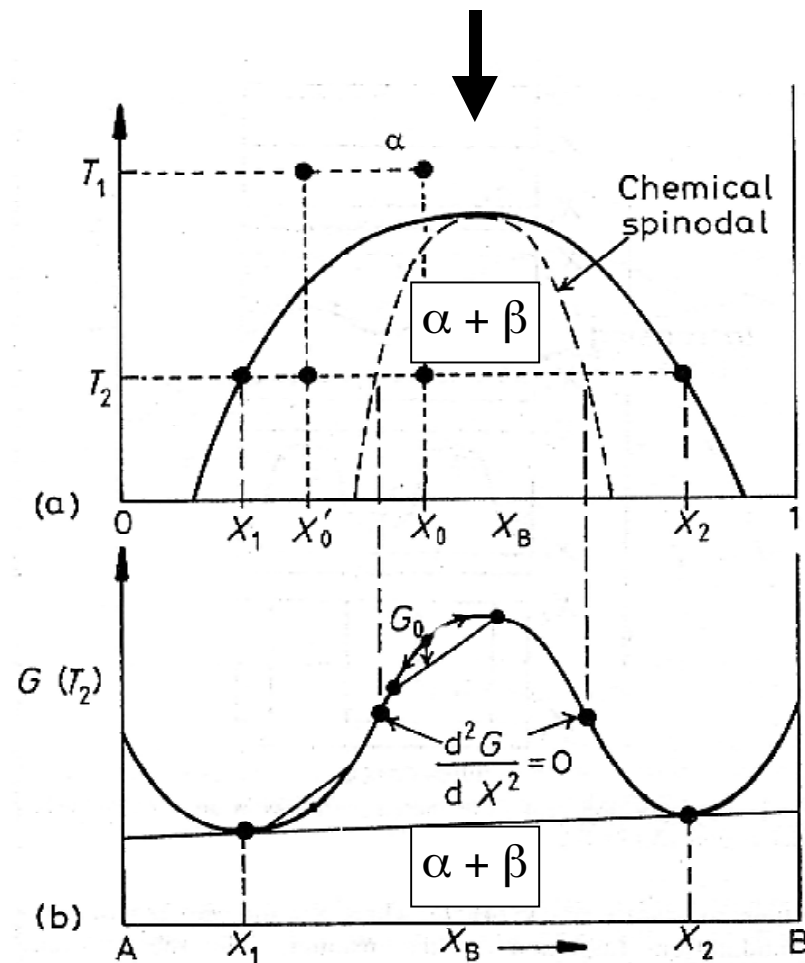
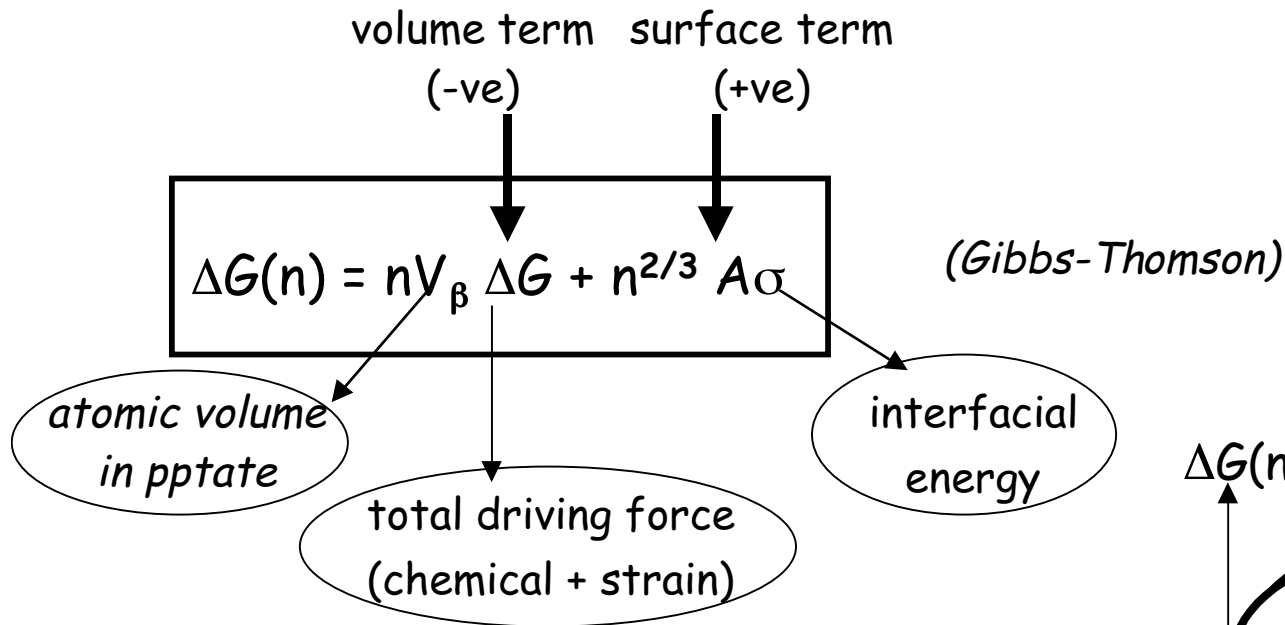


Fig. 5.38 Alloys between the spinodal points are unstable and can decompose into two coherent phases α_1 and α_2 without overcoming an activation energy barrier. Alloys between the coherent miscibility gaps and the spinodal are metastable and can decompose only after nucleation of the other phase.

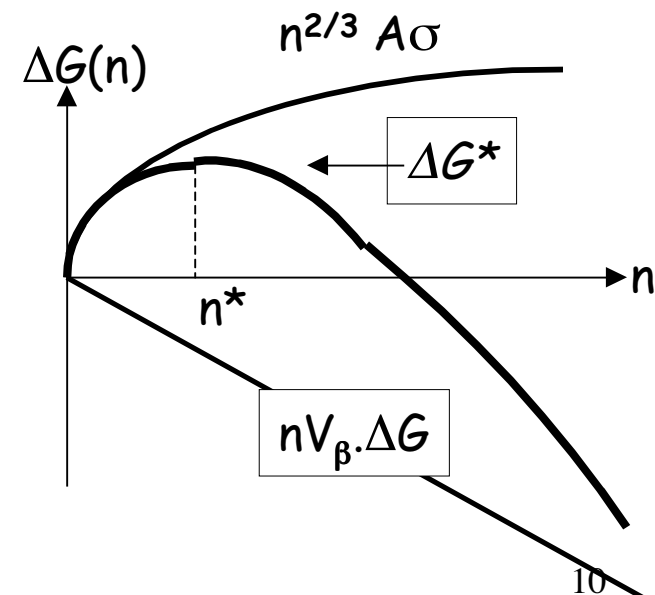
Thermodynamics: phase separation

Classical nucleation theory :

Formation of β -phase spherical cluster (n atoms) in α -phase :

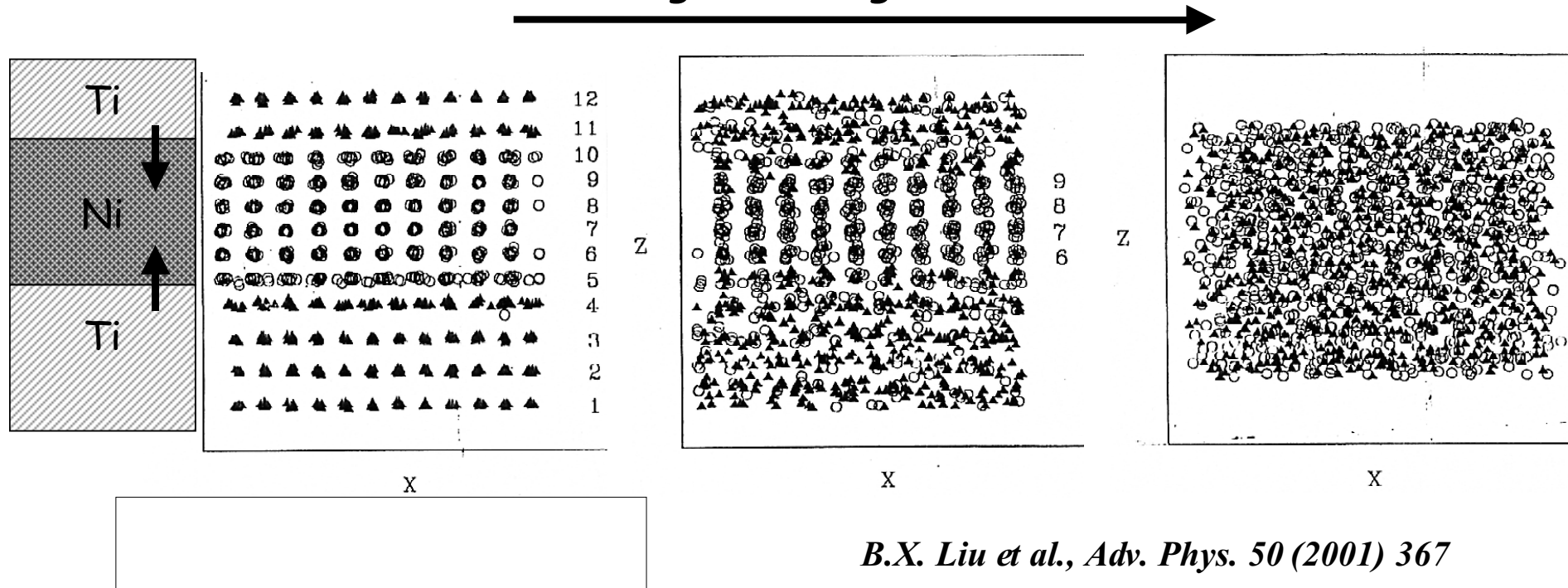


Rate theory :
Demixing kinetics $\Rightarrow J_{\alpha} - \exp(-\Delta G^*/kT)$

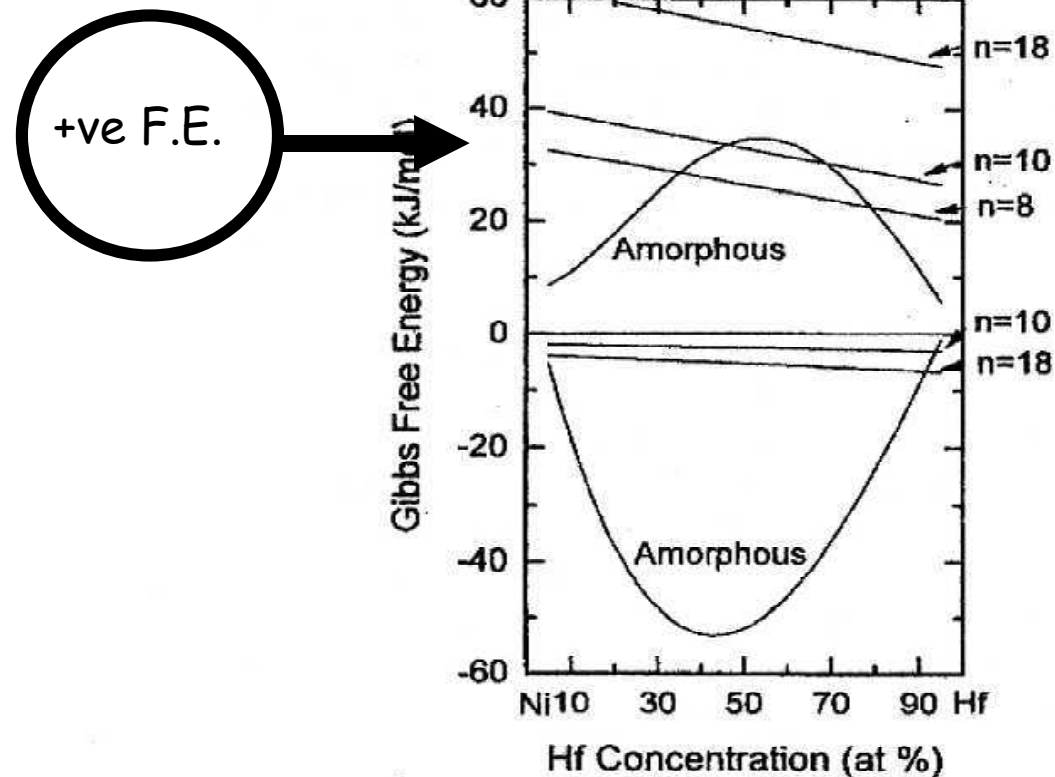


Interfacial growth of amorphous phase = $f(T)$

*Molecular Dynamics simulation
increasing annealing times @ constant T*



Rôle of interfacial energy

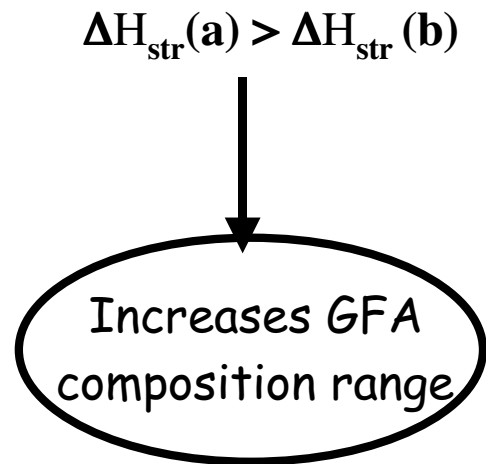


Calculated free energy diagram of the Nb–Y and Hf–Hf systems, showing different effects of interfacial free energy on alloy phase formation in the positive and negative heat of formation systems. n stands for the number of interfaces in the multilayers.

B.X. Liu et al., Adv. Phys. 50 (2001) 367

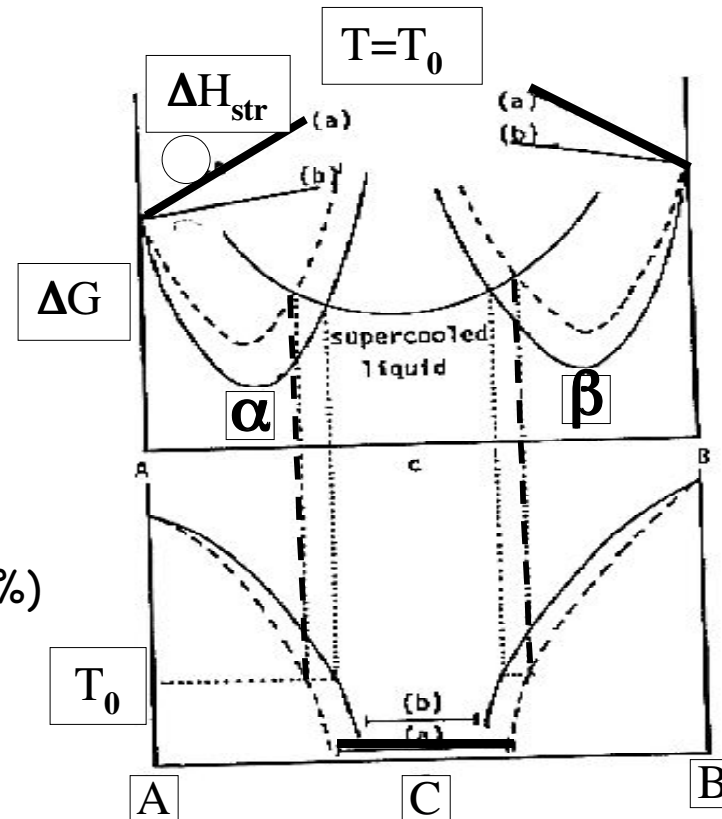
Thermodynamics of amorphous systems

Enhancing GFA : effect of **strain**



Rôle of

- component atomic size differences (>12%)
- interface mismatch (esp. @ nm scale!)



Critical parameters for
Glass Forming Ability (GFA)

- Volume free energy
- Competition between \neq structures
- Surface / interface energy
- Strain

Thermodynamics of amorphous systems

Enhancement of **G**lass **F**orming **A**bility (**GFA**) near deep eutectic

Presence of several phases with small ΔG differences \Rightarrow competing structures
(+ strain) favors amorphous phase formation

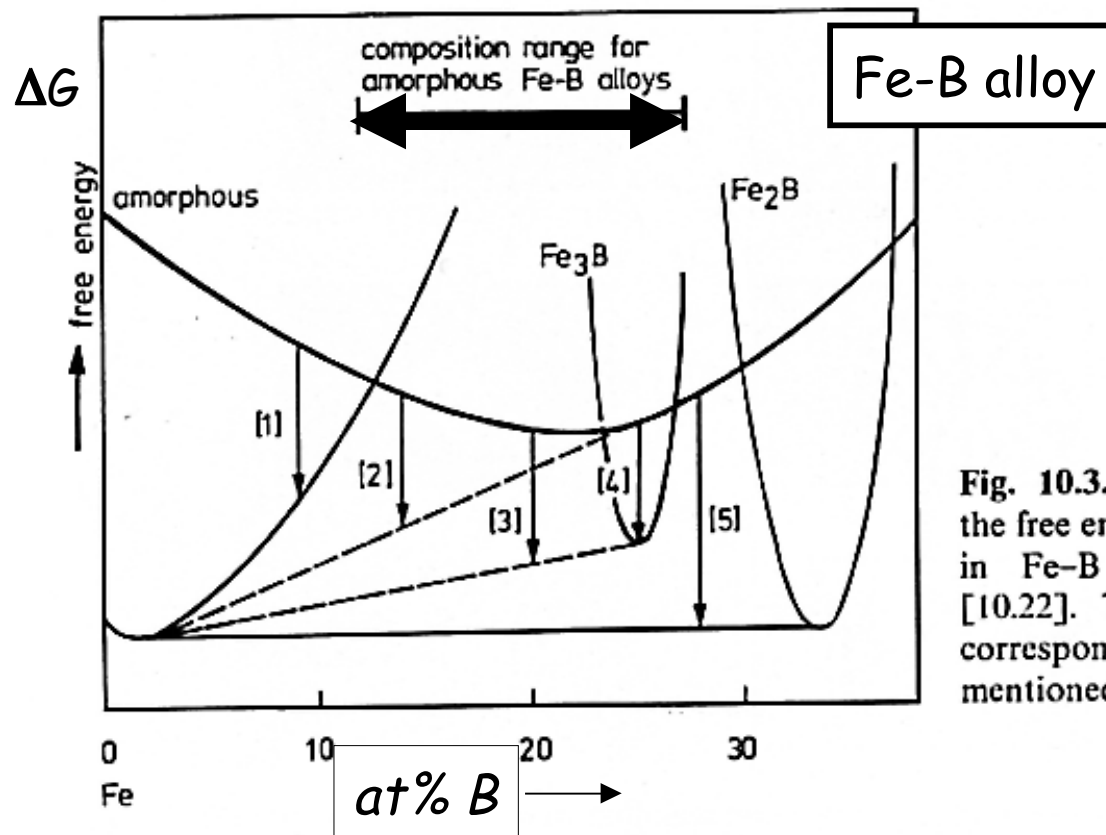


Fig. 10.3. Hypothetical diagram of the free energy for the various phases in Fe-B alloys vs concentration [10.22]. The numbers [1] to [5] correspond to crystallization reactions mentioned in the text

" Thermokinetics " of amorphous systems

Enhancing GFA: competing phases

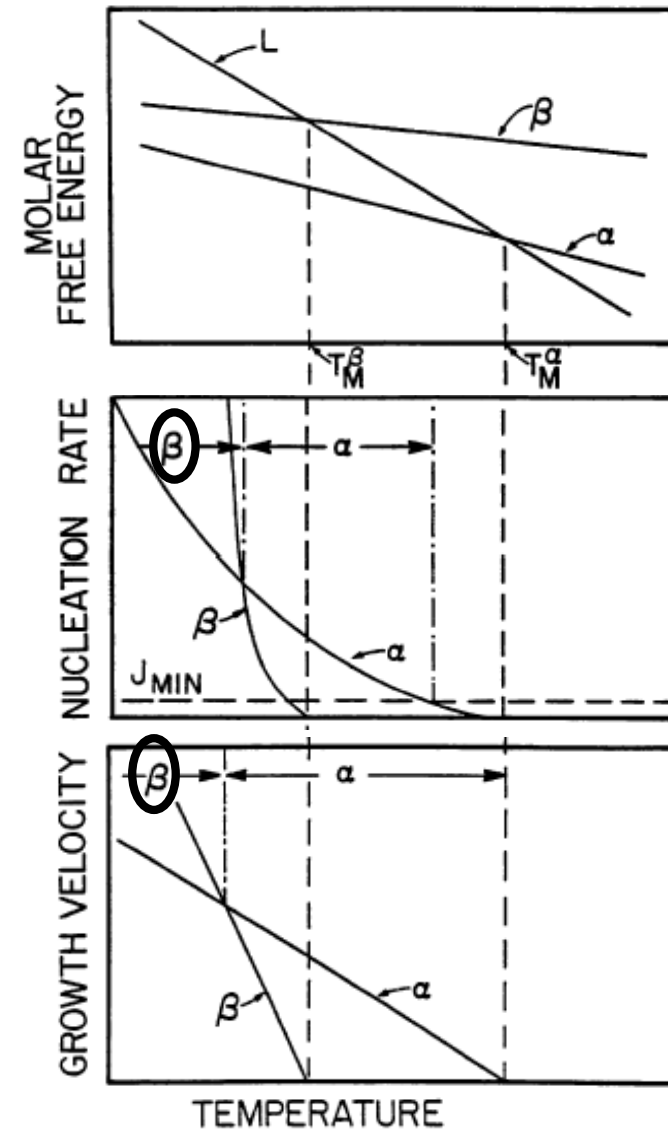
α -phase thermodynamically
most stable @ $T < T_M^\alpha$

effect of driving force
for nucleation @ low T

Rates $J_n \propto \exp(-\Delta E_a/kT)$

effect of driving force
for growth @ low T

Rates $J_g \propto \exp(-\Delta G^*/kT)$



Critical parameters for
Glass Forming Ability (GFA)

- Volume free energy
- Competition between \neq structures
- Surface / interface energy
- Strain
- Relative nucleation and growth speeds

Pre-irradiation conclusions

1. Free energy vs. kinetics -

- FE considerations => • GFA in deep eutectics
 - rôle of \neq phase structures & strain
 - nucleation & growth of *a* vs. *c* phase
 - GFA when competing phases => "frustration"

Note : on microscopic scale, atomic (+ vacancy) motion $\Rightarrow V_p, E_B \dots$ (MD, KLMC...)

2. Rôle of chemical short range order (CSRO)

- competing CSROs => favor GFA
- CSRO in amorphous structures

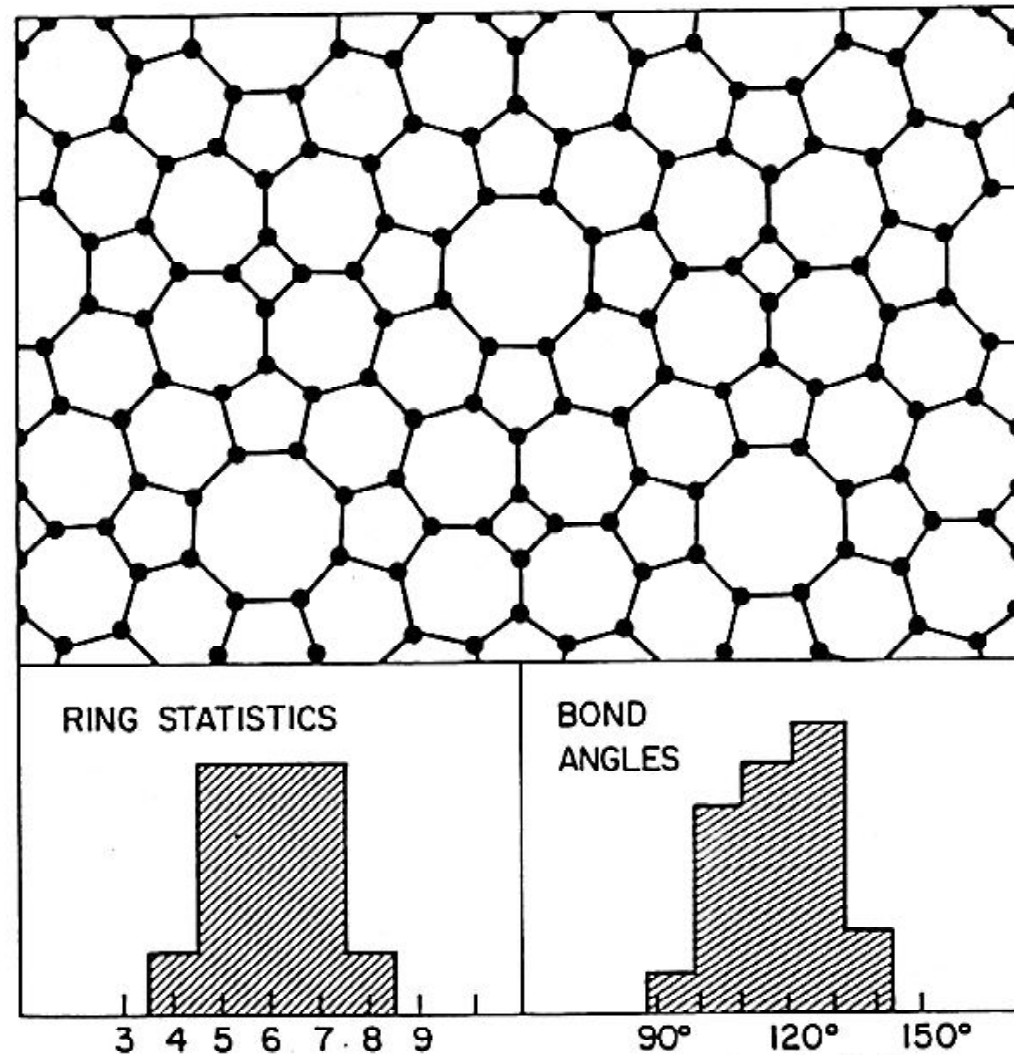
3. Transport and magnetic properties

- related to disorder **and** CSRO

Rôle of chemical short range order (CSRO)

"Amorphous" does **not** mean random !

- well-defined CSRO
- Distribution of bond angles \Rightarrow (+/-) MRO



Amorphous structure: an experimental definition

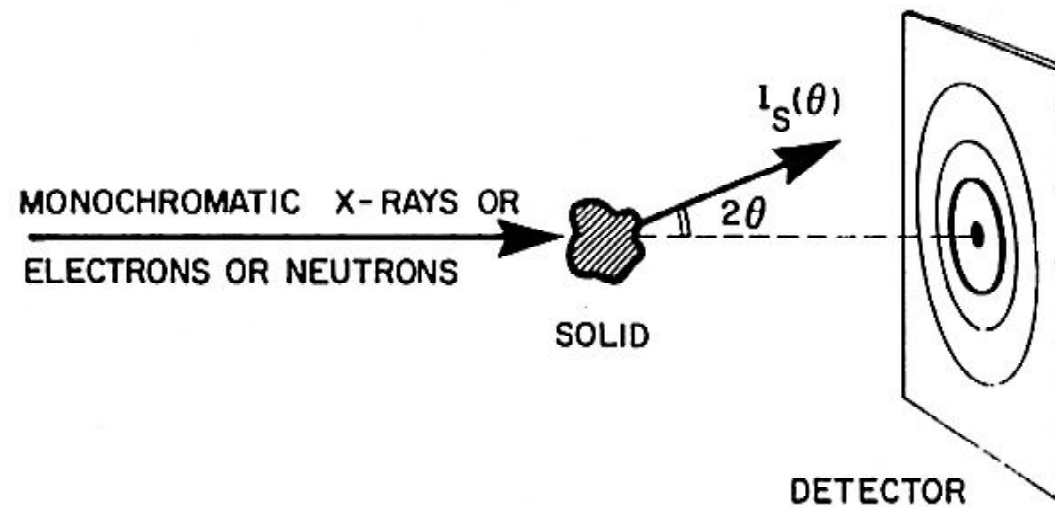


Figure 2.4 Basic geometry of structure-probing scattering experiments.

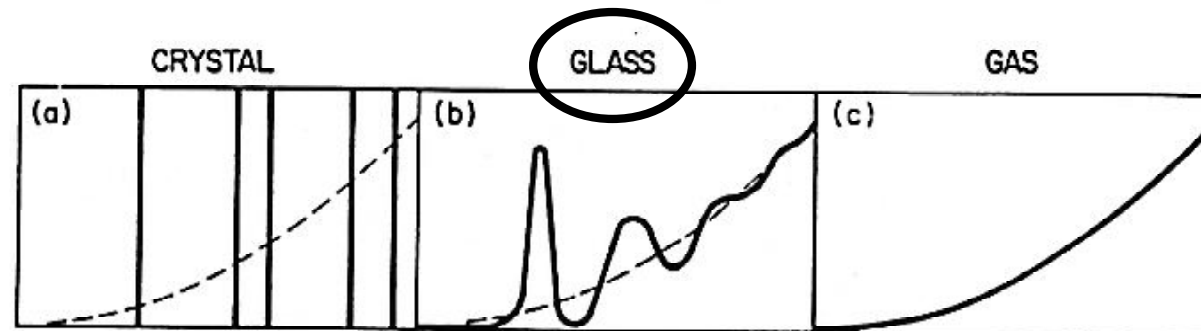
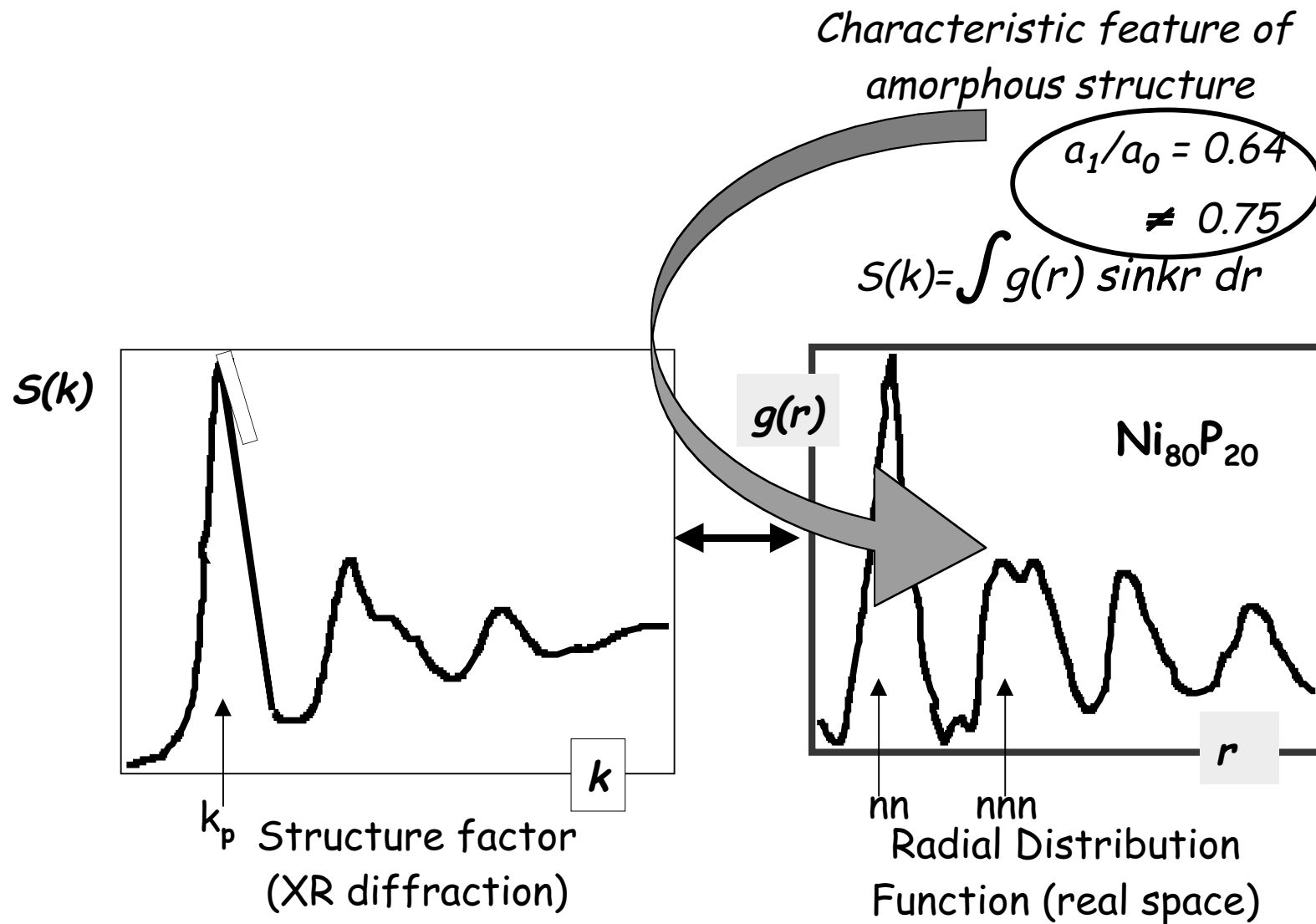


Figure 2.3 Schematic sketches of the radial distribution functions for (a) a crystalline solid, (b) an amorphous solid, and (c) a gas.

Amorphous structure



Amorphous structure

Probability of nn around A or B site

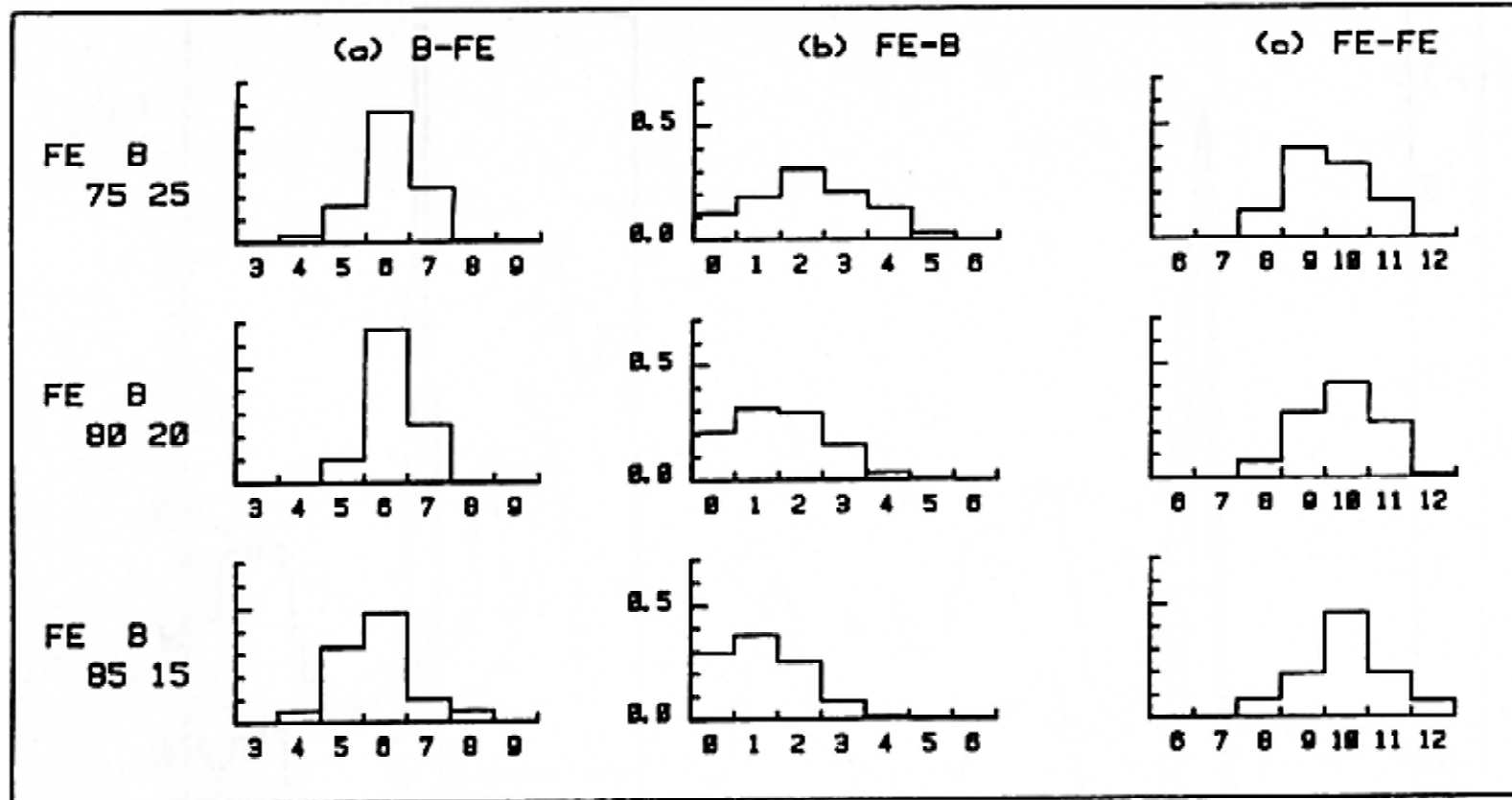
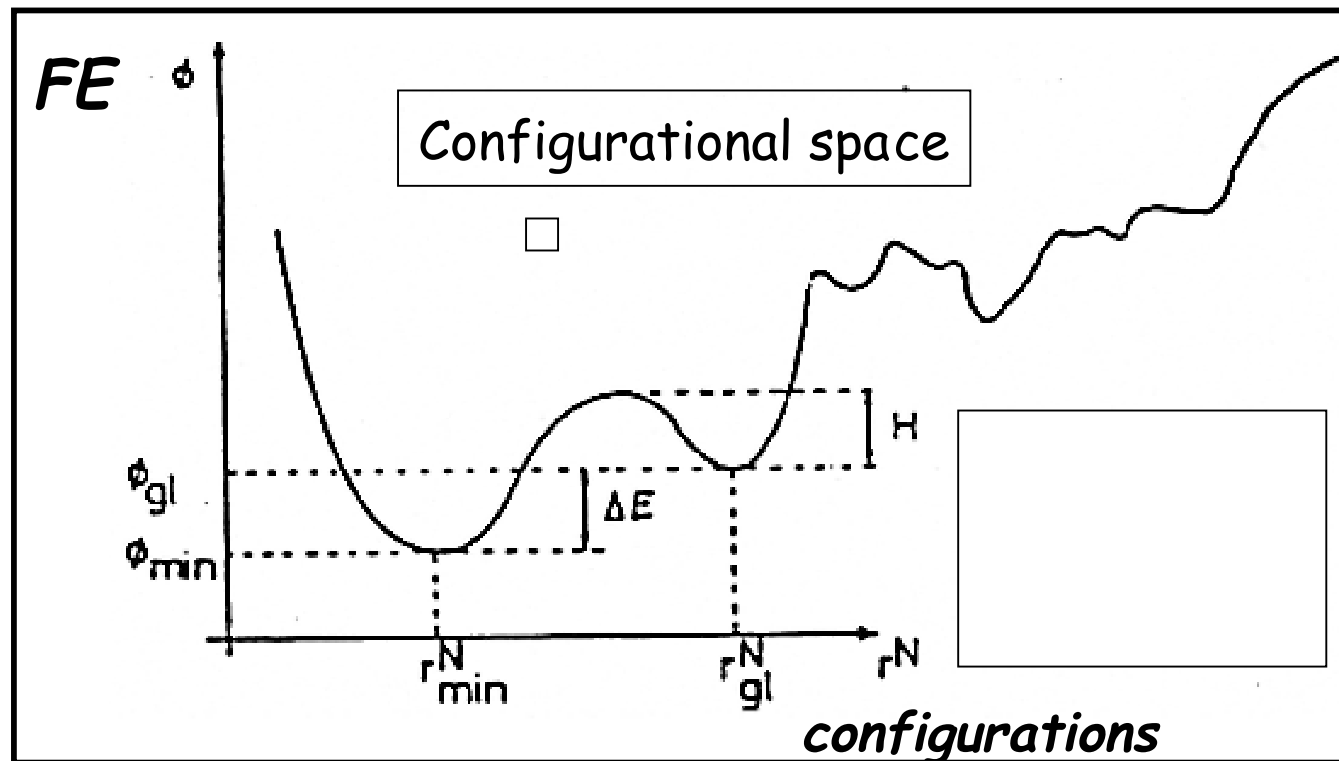


FIGURE 17. Probability distribution of the number of near-neighbors in models of Fe-B glass alloys. (a) Fe atoms around B atoms, (b) B atoms around Fe atoms, and (c) Fe atoms around Fe atoms. Near-neighbors are defined as atoms within the hard sphere touching distance $\pm 10\%$. (D. S. Boudreaux)

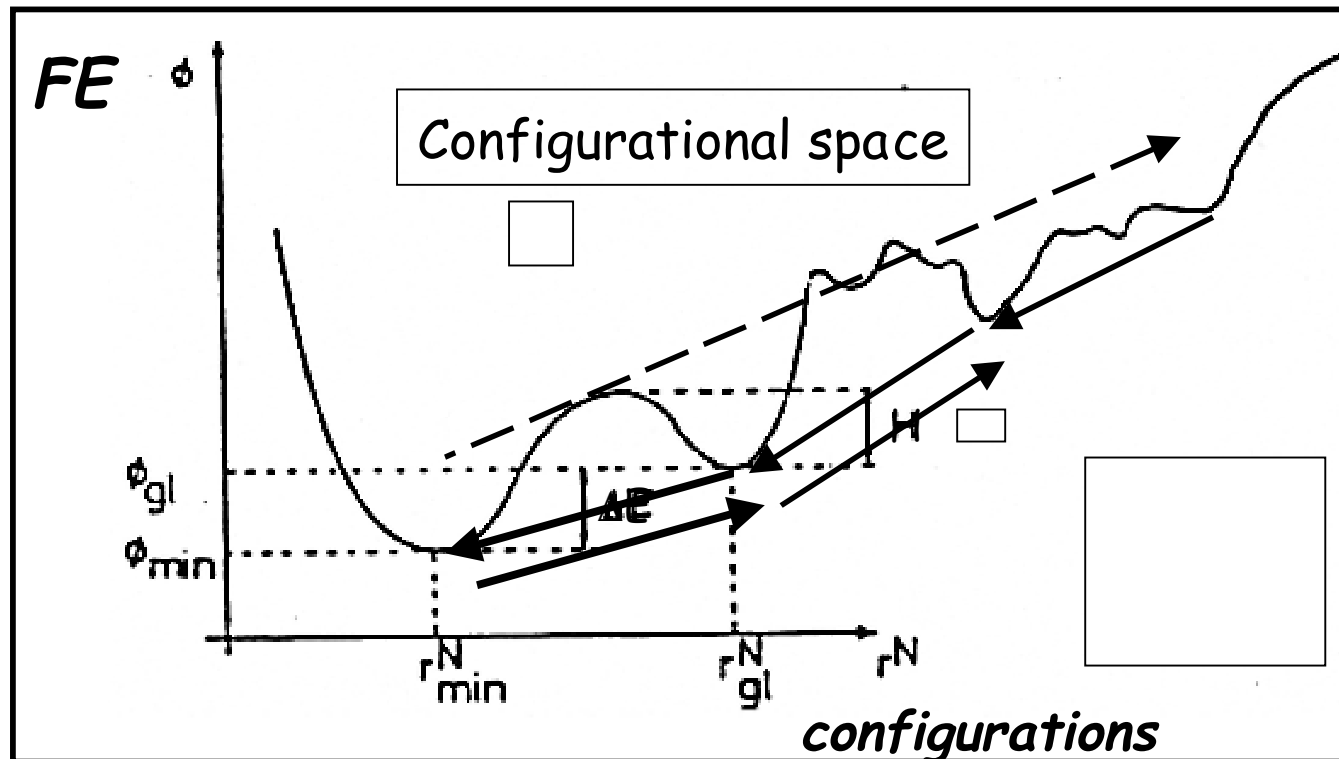
But as soon as one goes
from CSRO to
"Medium Range Order" (MRO)...

There are many different amorphous states ...



... and irradiation allows us to reach them

... and annealing or irradiation allows us to travel among them



(No phase transition !)

Pre-irradiation conclusions

1. Free energy vs. kinetics -

- FE considerations => • GFA in deep eutectics
 - rôle of \neq phase structures & strain
 - nucleation & growth of *a* vs. *c* phase
 - GFA when competing phases => "frustration"

To remember : Consider ΔG_0 (stable phase) vs. ΔG_1 (metastable phase)

- If $\Delta G_0 \gg \Delta G_1 \Rightarrow$ thermodynamics dominates
- If $\Delta G_0 \sim \Delta G_1 \Rightarrow$ kinetics determine phase formation

2. Twofold rôle of chemical short range order (CSRO)

- competing CSROs => favor GFA
- CSRO in amorphous structures

3. Transport and magnetic properties related to disorder and CSRO

Note : On a microscopic scale, **all effects** are related to atomic (+ vacancy) motion.

Effect of microstructure on transport properties

1. Basics :

- Boltzmann equation:

$$\rho \sim [(e^2 v_F^2 N(E_F) \tau)^{-1}]$$

- for liquid metals : ($\rho \neq$ random scattering) e^- scattering by correlated ion distribution \Rightarrow coherence of diffracted waves. mfp is determined by Faber-Ziman (weak scatt'g, pseudopotential approx.)

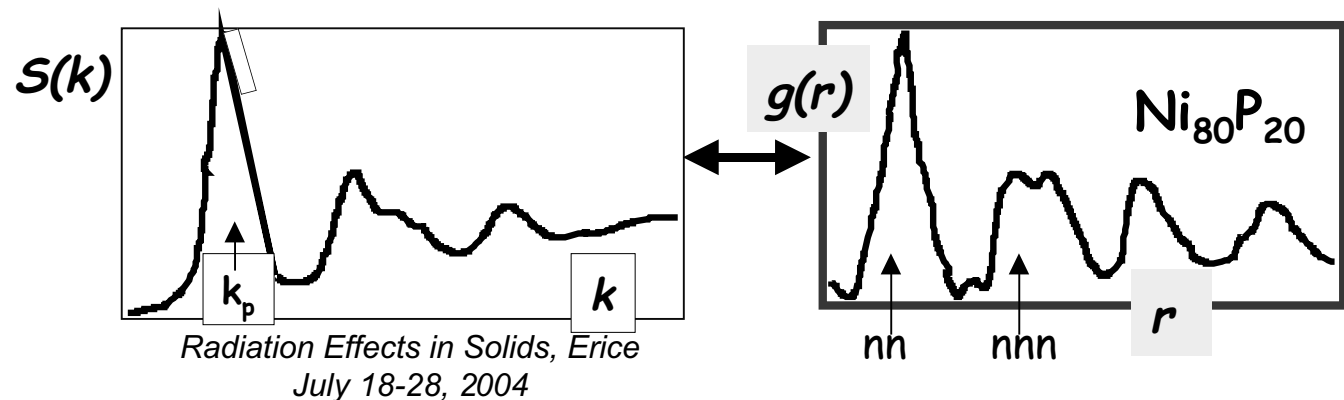
$$\tau^{-1} = 1/h \int (1 - \cos\theta) \left| \langle k | V | k' \rangle \right|^2 N(E_F)$$

with $V = \sum_i V(\underline{r} - \underline{R}_i) \Rightarrow \langle k | V | k' \rangle = [V(k)/N] \sum_i \exp(i\mathbf{k} \cdot \underline{R}_i)$

Now $(1/N) \left| \langle k | \sum_i \exp(i\mathbf{k} \cdot \underline{R}_i) | k' \rangle \right|^2 = S(\mathbf{k})$ structure factor of liquid

$$\Rightarrow \rho \sim [(e^2 v_F^2 N)^{-1} \int [V(k)]^2 S(k)]^4 (k/2k_F)^3 d(k/k_F)$$

Importance of CSRO !!



Pre-irradiation conclusions

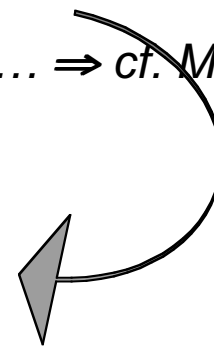
1. Free energy vs. kinetics -

- FE considerations =>
 - GFA in deep eutectics
 - rôle of \neq phase structures & strain
 - nucleation & growth of **a** vs. **c** phase
 - GFA when competing phases => "frustration"

Microscopic scale: related to atomic (+ vacancy) motion $\rightarrow V_P, E_B, \dots \Rightarrow$ cf. MD, KLMC...

2. Twofold rôle of chemical short range order (CSRO)

- competing CSROs => favor GFA
- CSRO in amorphous structures



3. Transport and magnetic properties related to

- disorder => aperiodic potentials; mfp \sim nm
- hence **CSRO** determines details of conductivity

Effect of irradiation/implantation

1. Free energy vs. kinetics -

- FE considerations => • GFA in deep eutectics
 - rôle of \neq phase structures & strain
 - nucleation & growth of *a* vs. *c* phase
 - GFA when competing phases => "frustration"

Modifications • **forced atomic mobility**

- phase diagram (radiation-enhanced diffusion)
- *(i) open system; (ii) far from equilibrium (ergodicity ?)*

2. Twofold rôle of chemical short range order (CSRO)

- competing CSROs => favor GFA
- CSRO in amorphous structures

Does irradiation (atomic displacements) modify **local stability** criteria ?

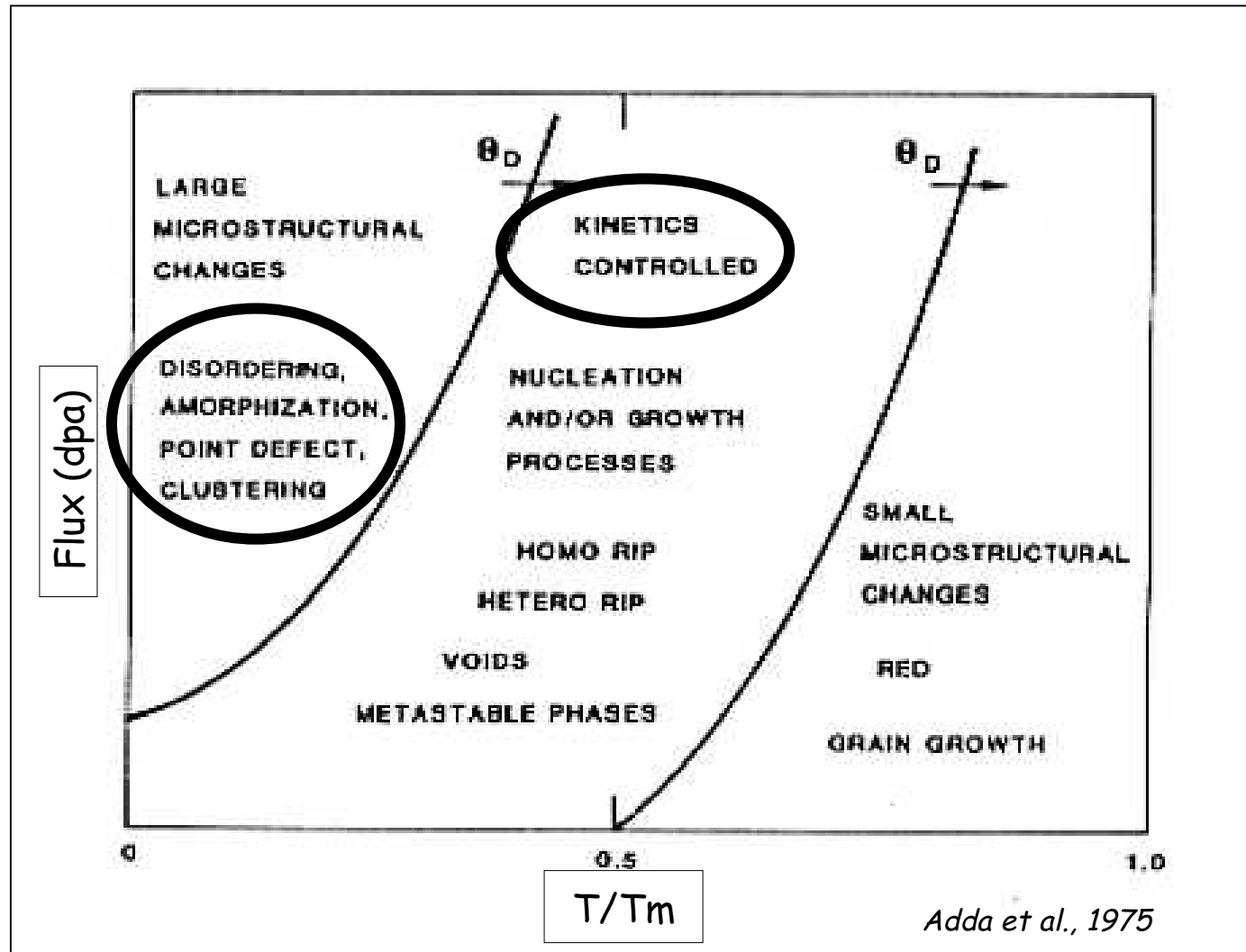
3. Transport and magnetic properties related to

- disorder => aperiodic potentials; mfp ~ nm
- hence CSRO determines details of conductivity

Sensitive to structure @ **nanoscale** (amorphization mechanism)

**Refer to thermodynamics
(equilibrium vs. nonequilibrium)**

Solids under irradiation



Open systems : irradiation and implantation

"open system" :
• stability does **not** require potential minimization
• there may be no unique equilibrium state

Characteristic features :
• irradiation : atomic mobility vs. temperature
• implantation : mobility + source term ($J_\alpha, \nabla c_\alpha$)

Control parameter:

• density of moving atoms
(defects ; defect sinks)

$c ? \nabla c ? (Dt)^{1/2} ?$

Ergodicity :
• a system is ergodic if ensemble average = time average
 $(1/N) \sum_1^N [y^{(k)}(t)] = \langle y(t) \rangle = \langle y \rangle \Rightarrow$ stationary system (A, B) goes through all intermediate states

• irradiation \Rightarrow **NON-ERGODIC**

Analysis of open ("driven") systems

G. Martin & P. Bellon, *Sol. State Phys.* **50**(1997)189

Mat. Sci. Forum **15-18**(1987)1337

J.W.Cahn, *Acta Met.* **9**(1961)795

Consider a 2-component system:

irradiation \rightarrow **competition** $\left\{ \begin{array}{l} \bullet \text{ random ballistic mixing} \rightarrow \text{instability} \\ \bullet \text{ irradiation-enhanced diffusion} \rightarrow \text{stability} \end{array} \right.$

Ballistic diffusion : $\mathcal{D}_B^* = c_2 \mathcal{D}_1 + c_1 \mathcal{D}_2$ where $\mathcal{D}_\alpha = \frac{1}{2} \Phi \sigma_r \langle r_\alpha^2 \rangle$
 irradiation flux \swarrow \searrow replacement cross-section

$\left\{ \begin{array}{l} \bullet \text{ ballistic interdiffusion flux } J_1^B = J_2^B = -\mathcal{D}_B^* N_v \partial c_2 / \partial x \\ \bullet \text{ irradiation-enhanced diffusion (} \sim \text{ JW Cahn) } J_{\text{diff}} = -M^* N_v [f'' \partial c_2 / \partial x - 2K \partial^3 c_2 / \partial x^3] \end{array} \right.$
 irradiation-enh. mobility \swarrow \searrow 2d deriv. FE \searrow gradient E
 where $M^* = [c(1-c)/kT] \mathcal{D}^* \rightarrow$ irradiation-enh. diffusion coeff.

Combine the 2 equations \rightarrow concentration profile that minimizes F.E.

(closed) $F\{c(x)\} = N_v \int \{f[c(x)] + K(dc/dx)^2\} dx \quad (\rightarrow \text{zero flux})$

(open) $\Psi\{c(x)\} = N_v \int \{\psi[c(x)] + K(dc/dx)^2\} dx \quad \rightarrow \quad \Psi \sim \text{F.E.}$

$$\boxed{\Psi\{c, T\} = f\{c, T'\}}$$

Open system under irradiation

G. Martin, Phys. Rev. 30 (1984) 1424

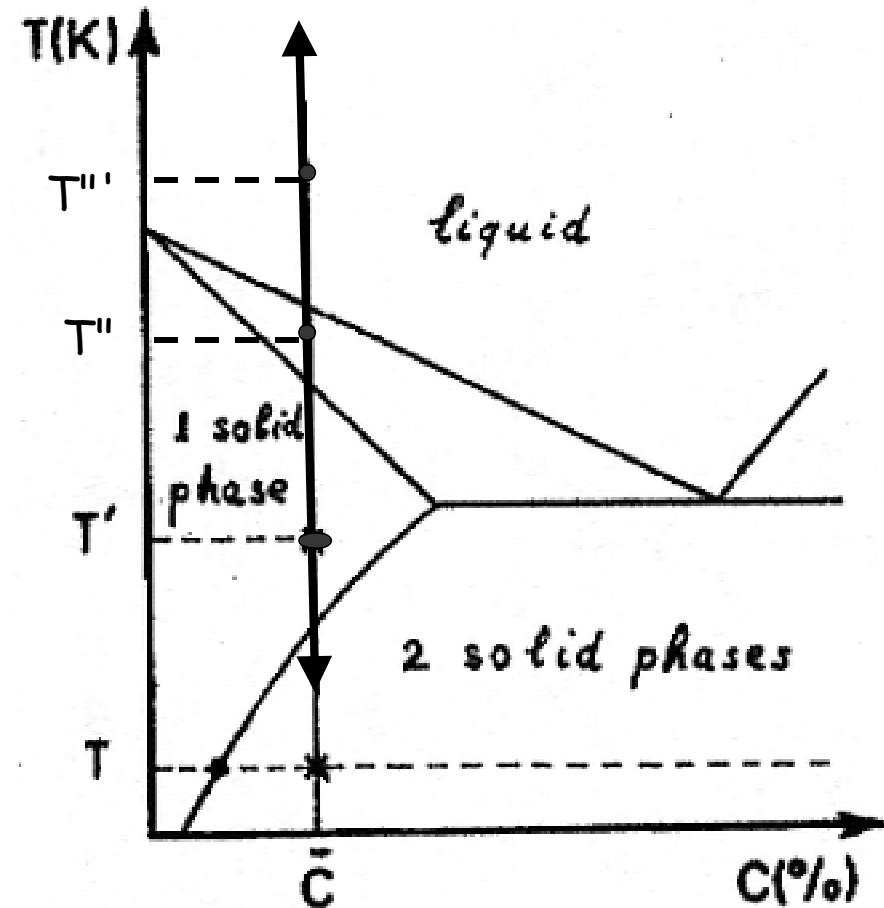
$$\begin{aligned} T' &= (1+\Delta)T \\ \Delta &= \mathcal{D}_B^*/D^* \end{aligned}$$

$$\Delta \propto \phi^{1/2} \exp(E_m^v/kT)$$

T' : precipitate dissolution

T'' : coexistence between amorphous & crystalline phases

T''' : amorphous phase



On a microscopic (mechanistic) scale...

Irradiation-induced disordering & amorphization are

- kinetic*
- non-ergodic*

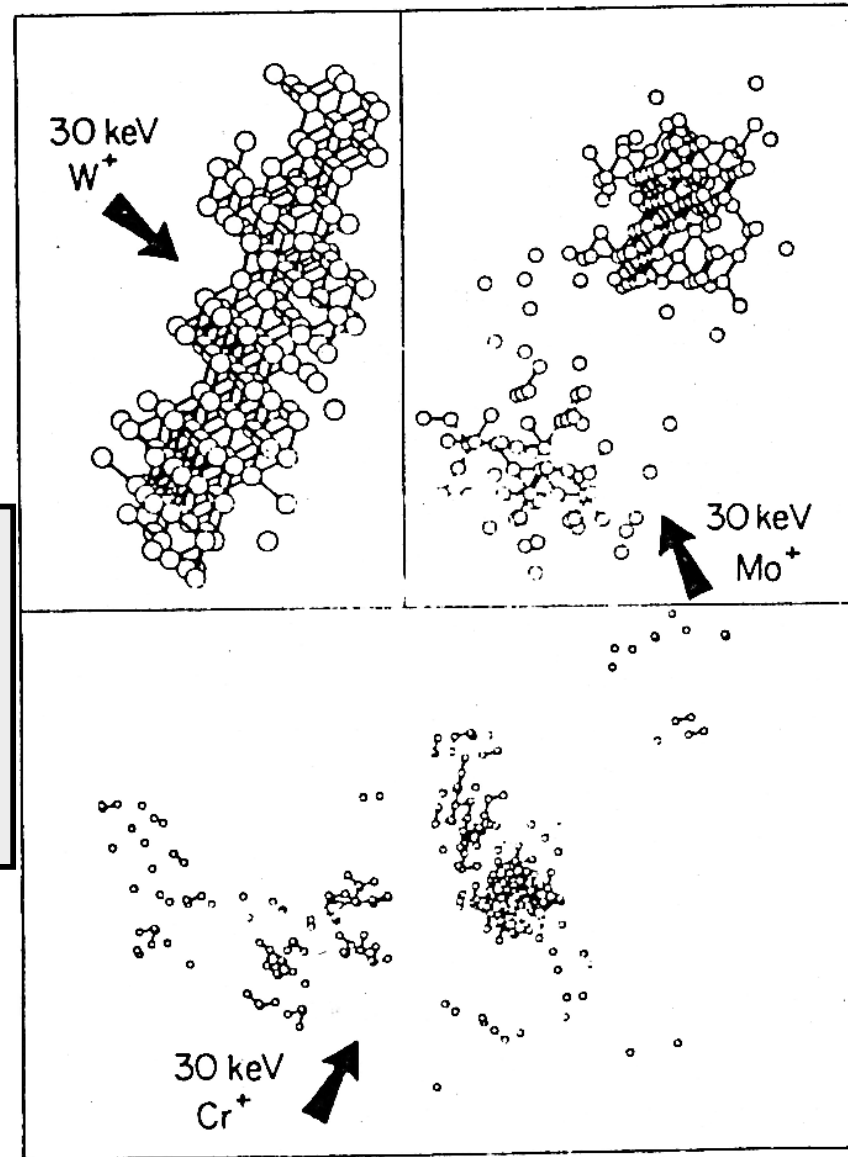
Displacement cascade residues in W @ RT



Degree of disordering/amorphization depends on

- deposited energy density
- nature of bonding
- defect recombinations

L. Wei & D.N. Seidman



Kinetic (collision-induced) amorphization

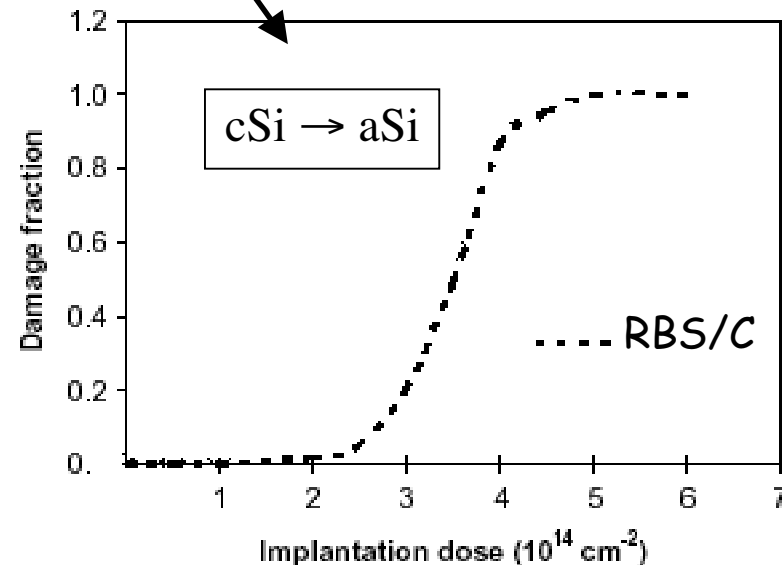
1. "Simple case 1 " (ex: c-Si → a-Si ?):

==== *Direct amorphization by each displacement cascade* ====

$$\Rightarrow \alpha(a/c) = [1 - \exp(-\sigma\Phi)] = P_1$$

F. F. Morehead & B. L. Crowder, Rad. Eff. Def. Sol. 6(1970)27

(alternatively, any Σ of defect mechanisms will do !)



(L.C. Feldman, T. Picraux, 1978)

Kinetic (collision-induced) amorphization

2. "Simple case 2" : collision-induced amorphization

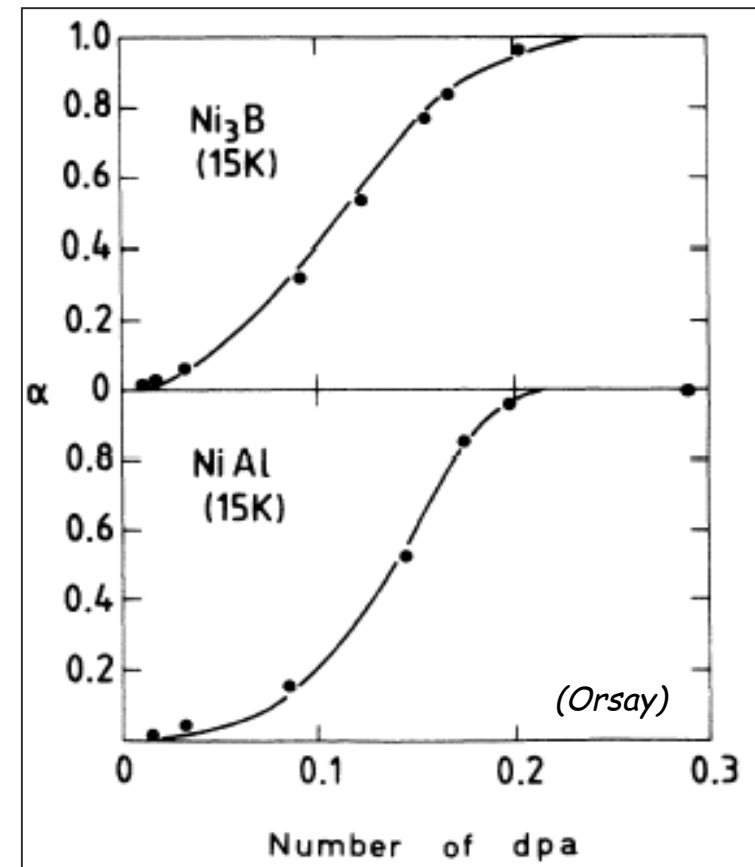
==== Need more than 1 impact to amorphize ====

Amorphization by displacement (cascade) overlap

$$\Rightarrow \alpha = \sum_{n_c}^{\infty} [n^n \exp(-\sigma\Phi t) / n!] = P_n$$

J. F. Gibbons, Proc. IEEE 60(1972)1062

ex: $c\text{-Ni}_3\text{B} \rightarrow \alpha\text{-Ni}_3\text{B}$
 $c\text{-NiAl} \rightarrow \alpha\text{-NiAl}$
(irrad. with keV D ions)



Kinetic (collision-induced) amorphization

3. Johnson-Mehl-Avrami mechanism: nucleation+growth+overlap

(→ involves duration of phase transformation)

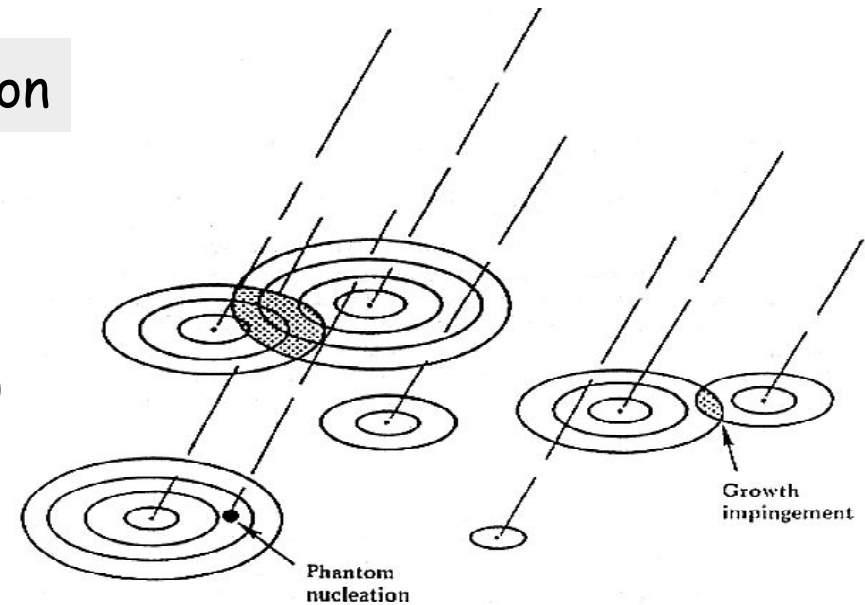
$$\Rightarrow \alpha = 1 - \exp(-Kr^n)$$

$$K \sim R_N R_G^3 \exp(-\Delta H_N + 3\Delta H_G)/kT$$

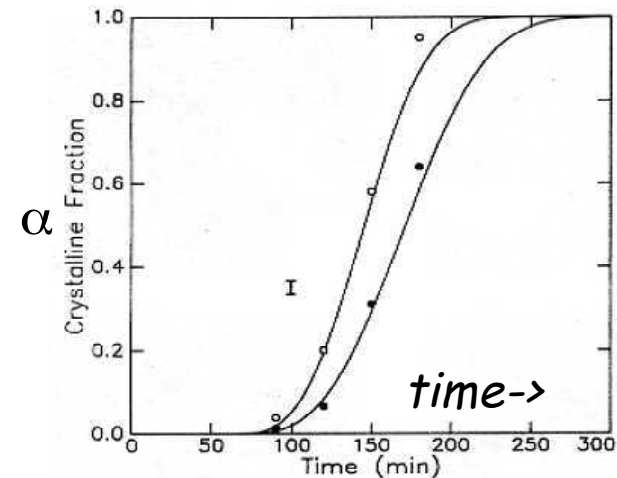
nucleation & growth rates

$$n = F(\text{nucl. \& growth modes}) = 1, \dots 4 \dots$$

Note! ~ same sigmoidal shape as #2



A schematic picture depicting the falling of raindrops on the surface of a pond. The overlapping of ripples (growth impingement) and a later raindrop falling within a ripple (phantom nucleation) are indicated.

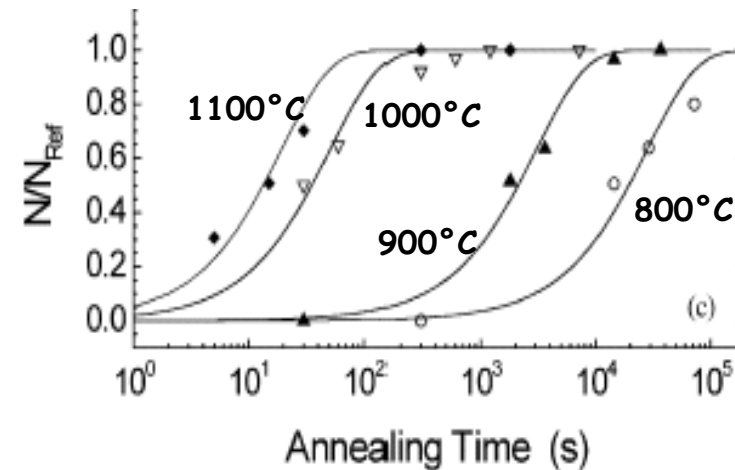
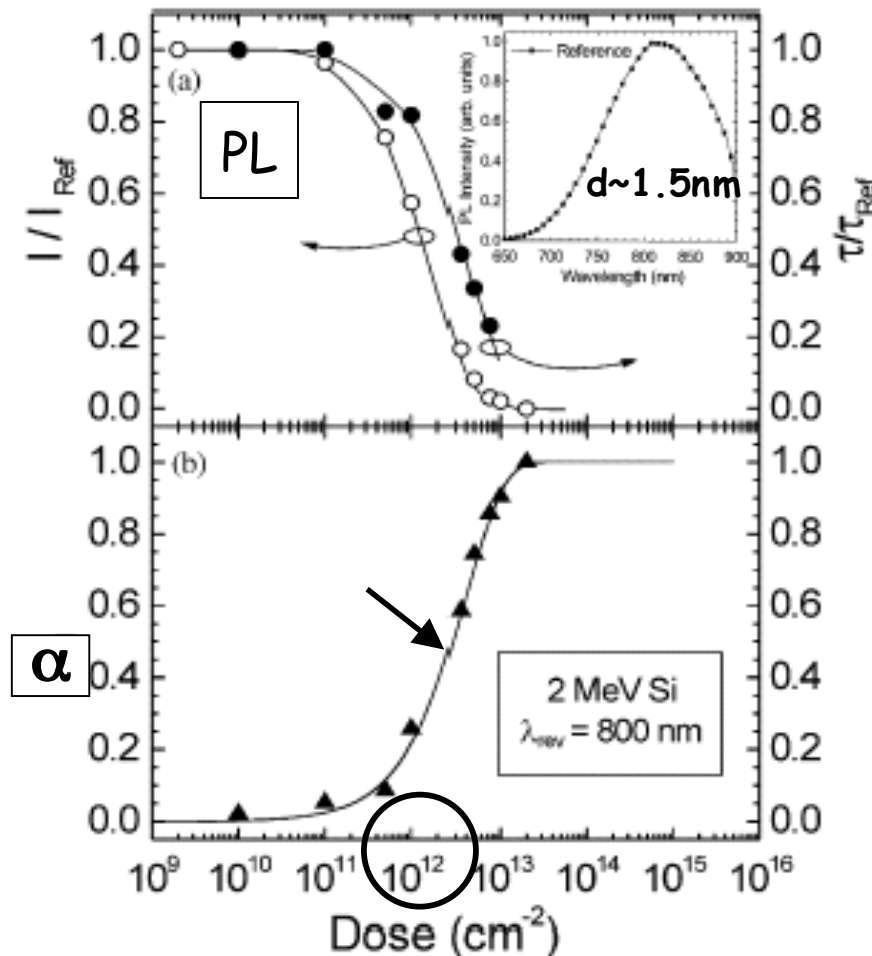
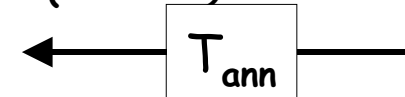


Importance : **Interface growth, where ballistic mixing competes with growth via diffusion/migration**

Rôle of interfaces: Amorphization & recrystallization of Si nanoclusters in SiO₂

D. Pacifici, G. Franzò et al, Physica E 16(2003)404

Irradiate by Si(2 MeV)



$$\frac{N}{N_{Ref}} = 1 - \exp\left(-\frac{t}{\tau_c}\right) \quad ??$$

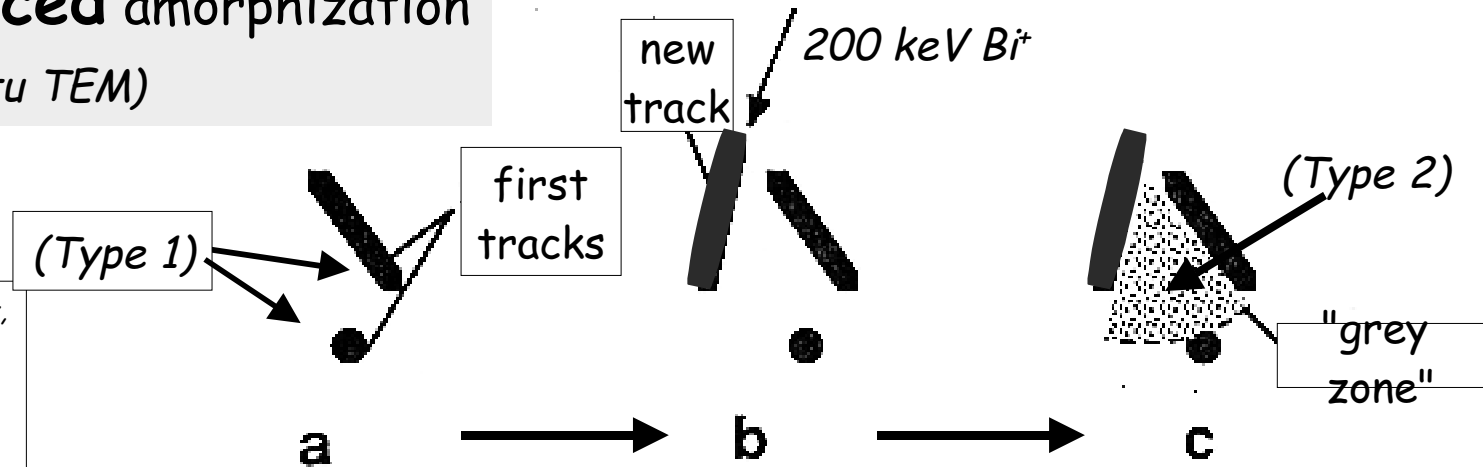
τ_c = recrystallization time

Nanoclusters vs. bulk:

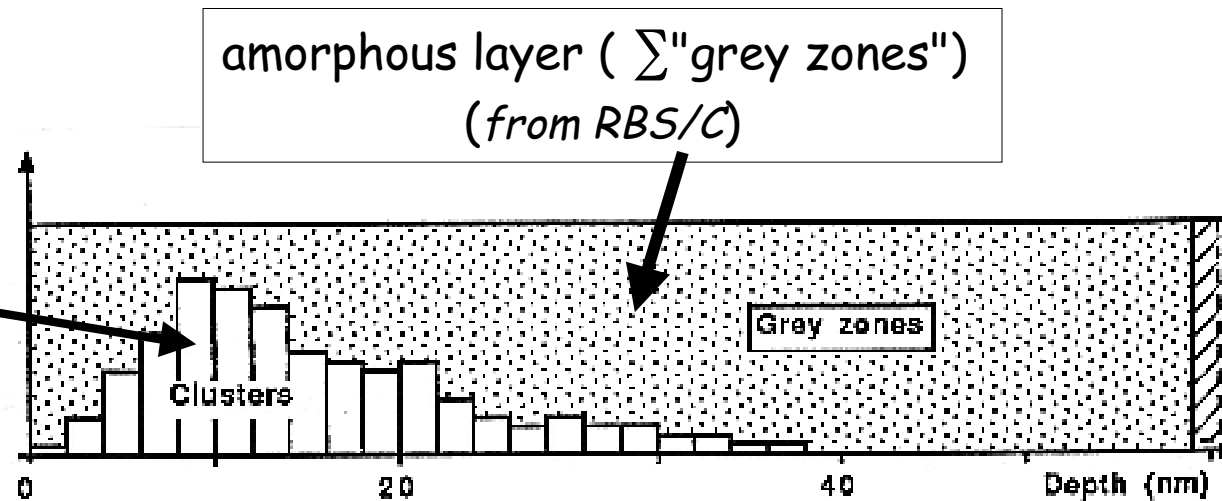
- amorphization faster
 - recrystallization slower
- Rôle of interface

Strain-induced amorphization in Si (from *in situ* TEM)

From H. Bernas, M.O. Ruault,
P. Zheng, "Crucial issues in
Semiconductor Materials &
Processing Technologies",
Kluwer (1992), p.459



strain-contrasted
amorphous clusters
= Type 1



Note: $T_{evol} \sim 1$ second @ RT ; $T_{ann}(\text{Type 1}) \sim 500K$; $T_{ann}(\text{Type 2}) \sim 800K$

Implantation =
Irradiation + compositional change:

Progressive c/a phase transformation



Building the amorphous structure from CSRO

Progressive amorphization of Ni-P via implantation

- ~ constant atomic displacement rate $\rightarrow \mathcal{D}_B^*/D^* = \text{cst. (total=100 dpa !)}$
- constant rate of P concentration increase

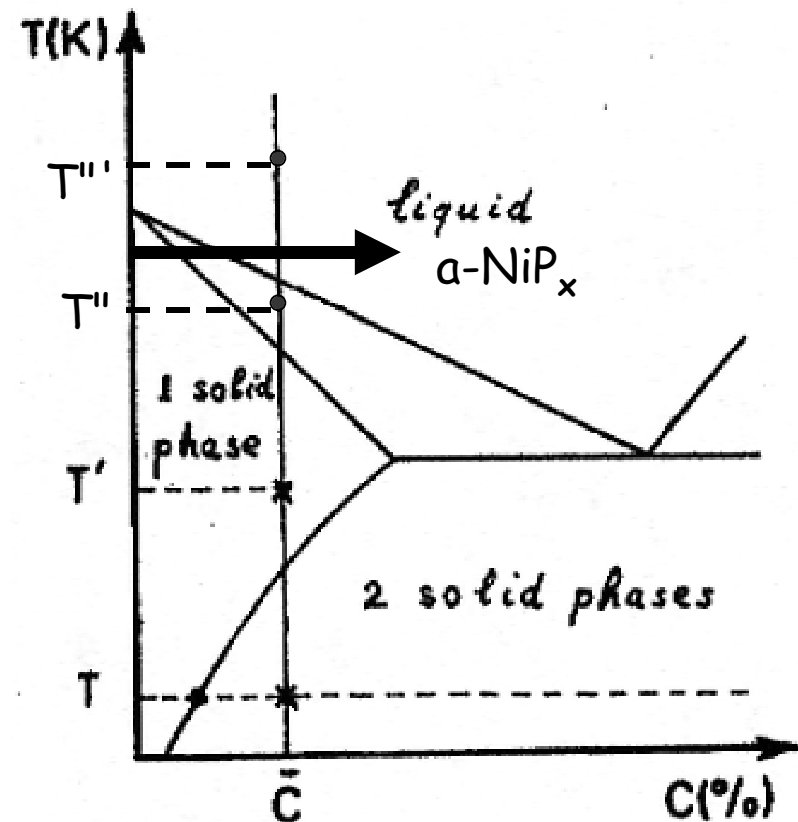
$$T' = (1 + \Delta)T$$

$$\Delta = \mathcal{D}_B^*/D^*$$

T' : precipitate dissolution

T'' : coexistence between amorphous & crystalline phases

T''' : amorphous phase



Progressive amorphization of Ni-P via implantation (2)

Questions

- @ low T "basic" amorphization mechanism:
kinetics > TD : rôle of GFA ?
- @ higher T , competition: kinetics vs. growth
- new properties of the implantation-induced amorphous state ?

The experiments

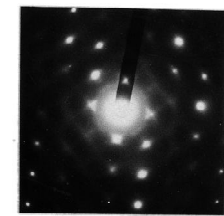
- Progressive implantation of P in Ni single crystals
- in situ TEM @ 15K & 300K: Diffraction studies
- RBS/C @ 80K & 300K: dechanneling studies

Results

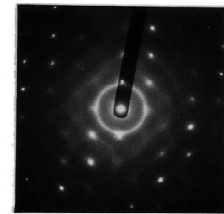
- 1 phase (Ni+ ϵ P) \rightarrow 2 phases (c -Ni + a -Ni_{0.88}P_{0.12}) \rightarrow 1 phase (a -Ni_{0.85}P_{0.15})
- "building block" of a -NiP_x: the stable amorphous cluster
- confirmation via magnetic properties

Complementary information from in situ TEM: Strain

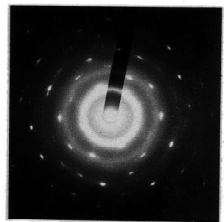
Amorphization of Ni-P_x @ 80K



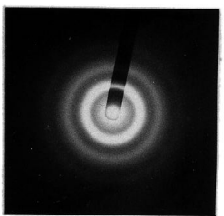
(e)



(f)



(g)



(h)

Ni/P⁺50keV

Phase Transformations-3

H. Bernas

In situ
TEM

RBS/C

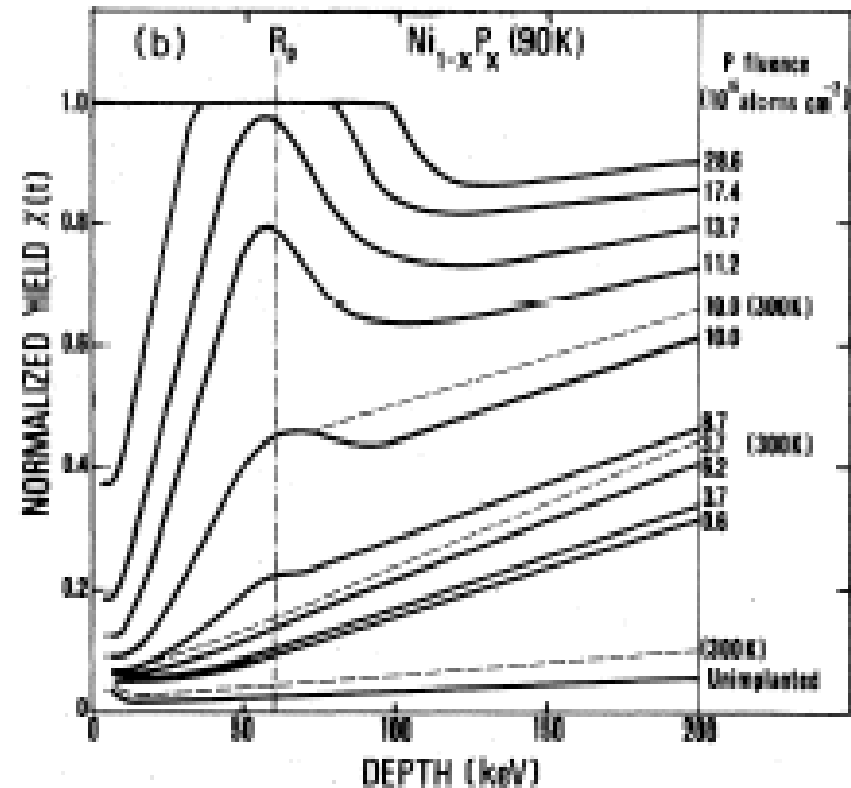


FIG. 4. Depth dependence of the [100]-aligned RBS yields normalized to the random yields for P-implanted Ni single crystal at [] LNT. Phosphorus-ion-implantation energy, 125 keV; analyzing particles, 380-keV ⁴He ions.

M. Schack, MO Ruault et al; C. Cohen et al., PRB (1986)

Amorphization of Ni-P_x @ 80K: the amorphization mechanism

From *TEM* & *RBS/C*, assume heterogeneous compound :

- amorphous clusters + crystalline host
- ➡ • atomic displacements play ~ no rôle @ >1 dpa
- amorphous clusters are
 - stabilized by a minimum amount of P atoms N_c
 - all identical
 - irradiation → relaxation

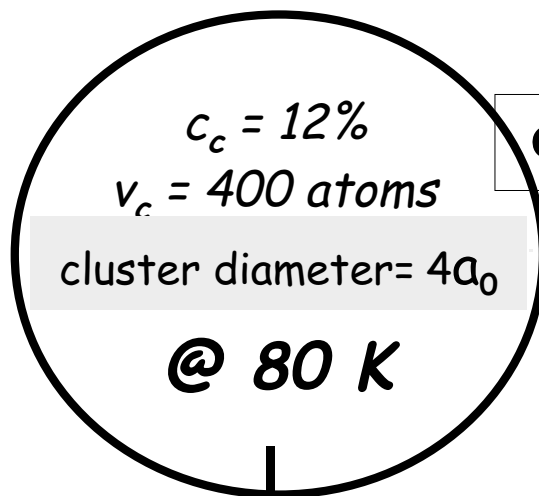
⇒ • macroscopically, nonequilibrium system
• microscopically, minimize configurational energy via clustering

- in implanted profile, divide volume → v_c ; ave.nbr. P atoms = \underline{n} ,
probability distribn. is $P(n) = (\underline{n})^n \exp(-\underline{n})/n!$
- ⇒ amorphous fraction is

$$\alpha = \sum_{N_c}^{\infty} [(\underline{n})^n \exp(-\underline{n})/n!]$$

*Building up amorphous NiP_x by
P implantation in Ni*

*From RBS/C
(confirmed by TEM)*



• $c_c = 19\%$
 (deep eutectic)
 • include SRO growth
@ 300K

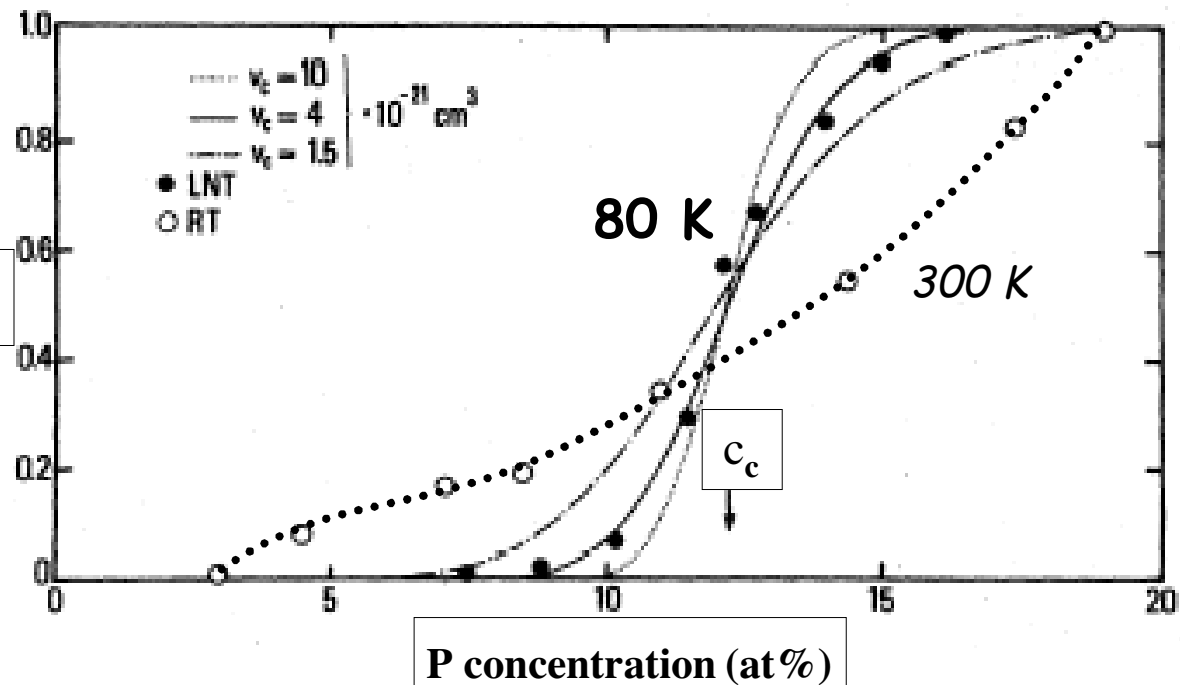
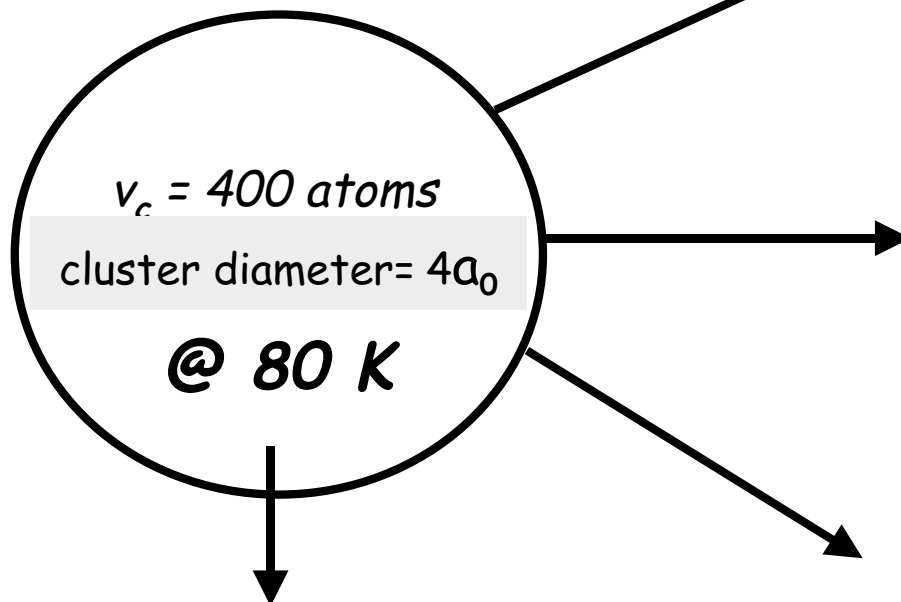


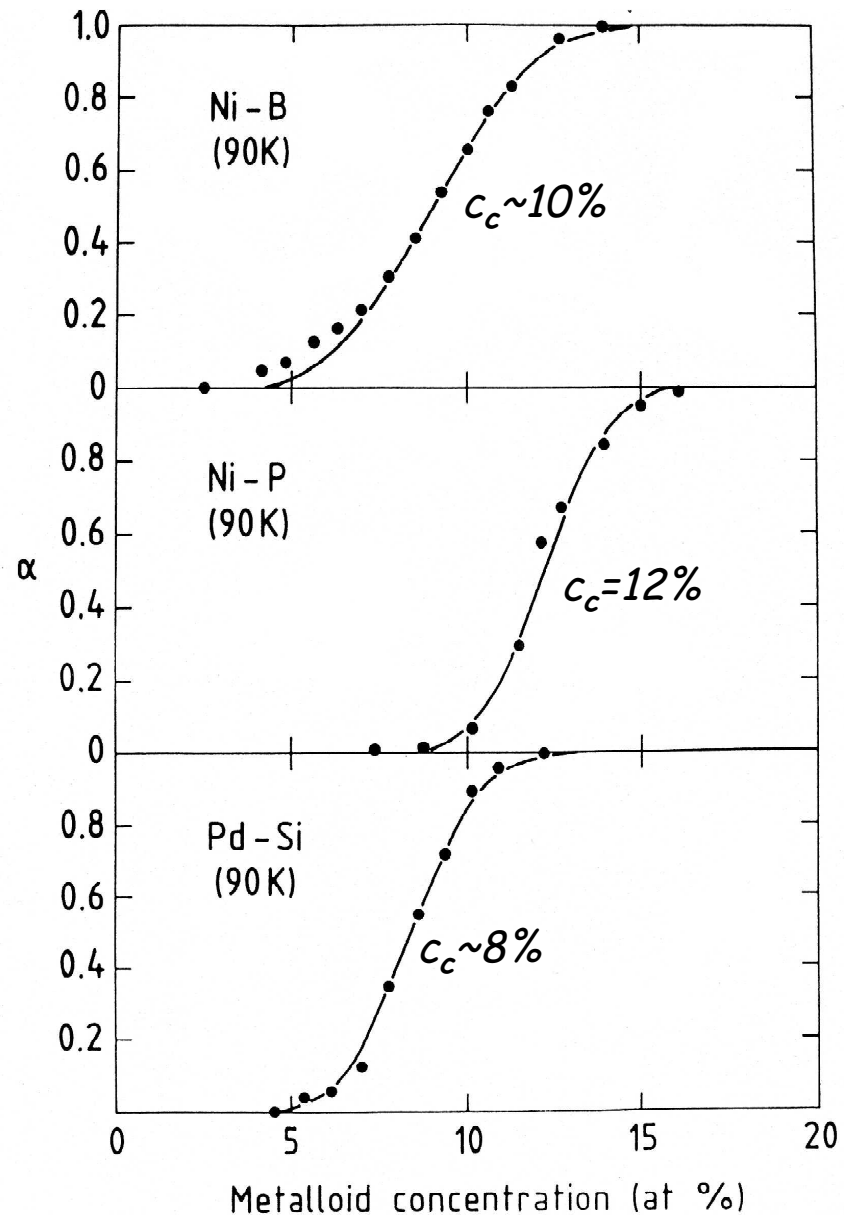
FIG. 8. P-concentration dependence of the amorphous fraction α for LNT (solid circles) and RT (open circles) implants. The solid line represents the best fit to the LNT experimental data using Eq. (4). Dashed lines are calculated, using Eq. (4), with different elementary volumes v_c of the amorphous clusters

*Building up amorphous NiP_x by
P implantation in Ni*

*From RBS/C
(confirmed by TEM)*

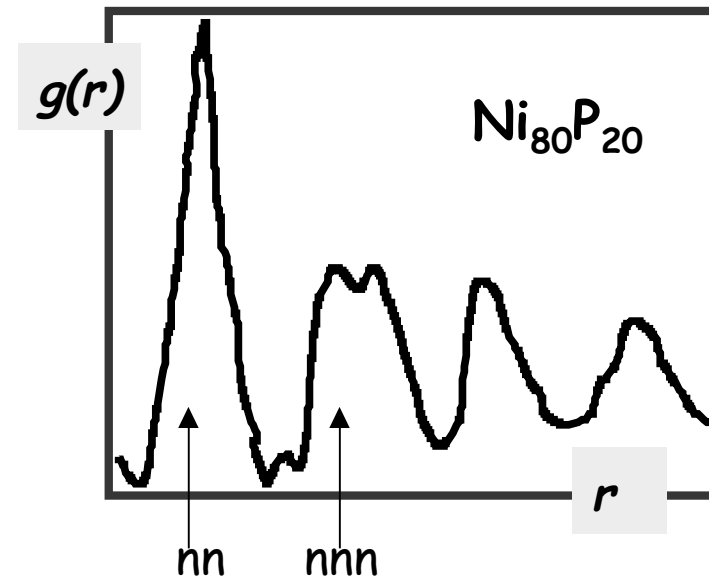
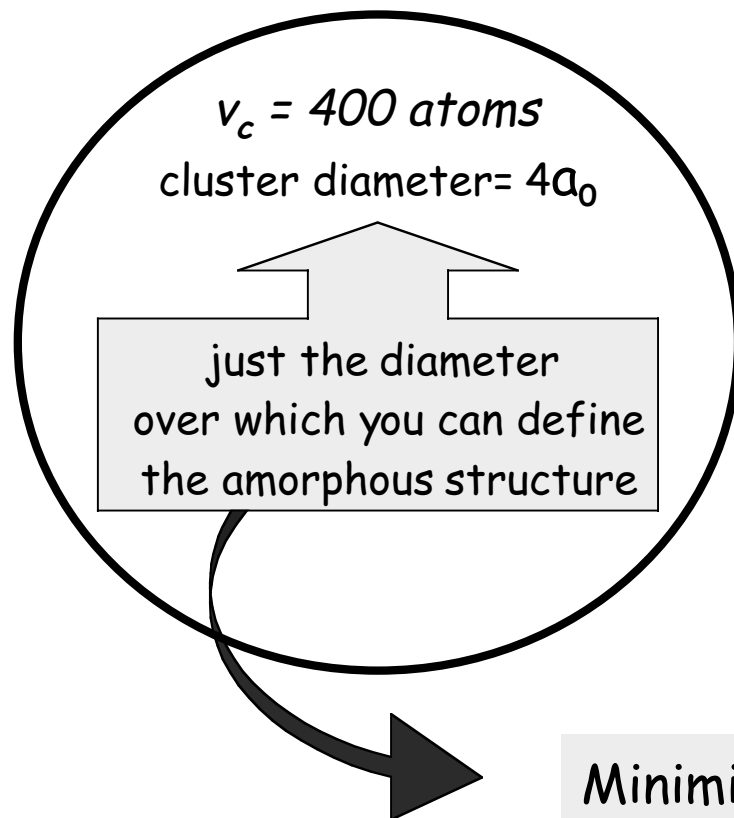


- $c_{\text{Max}} \sim 19\%$
(deep eutectic)
- include short-range growth



L. Thomé, A. Audouard et al., PRB (1987-88)

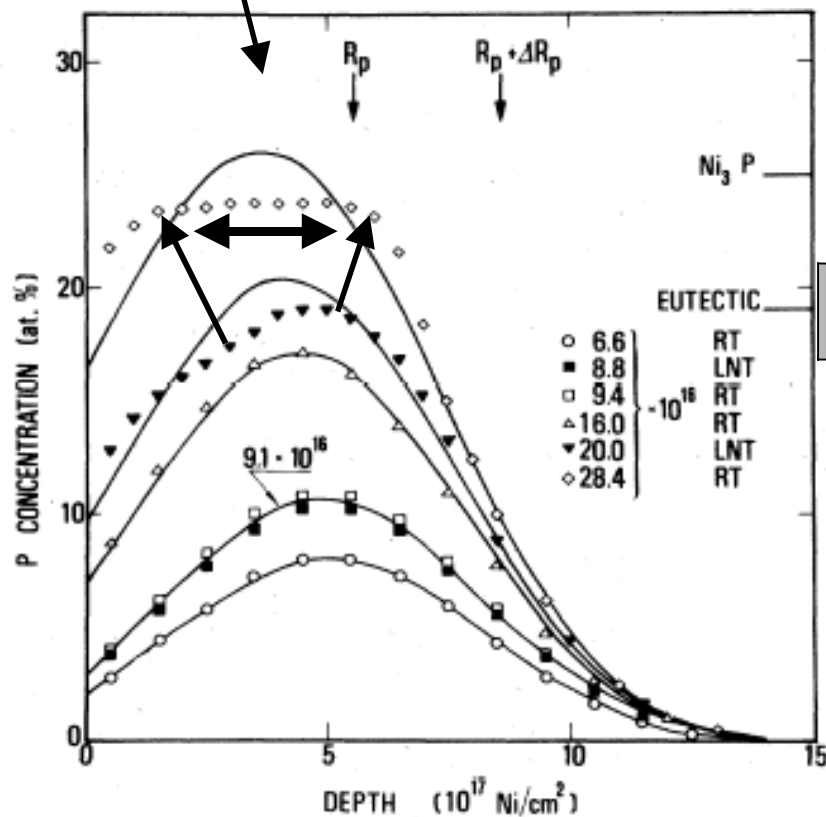
Stability criterion for building the amorphous phase



H. Bernas et al., (1986)

Amorphization mechanism: 80 K vs. 300 K

Short range P diffusion inside
the implantation profile @ 300K



Relaxation of
amorphous cluster
composition:

$P_{0.12} \rightarrow P_{0.20}$

deep eutectic

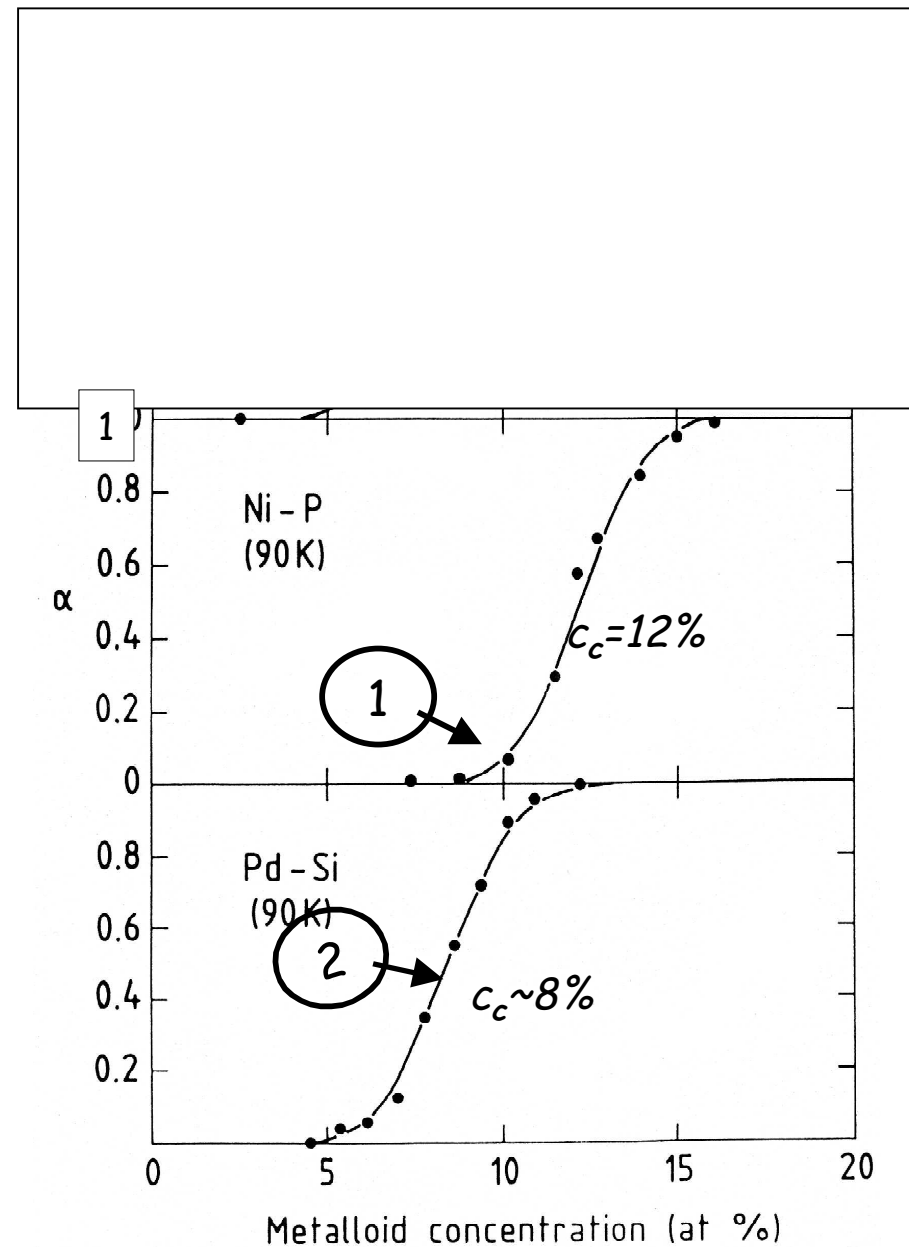
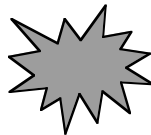
Kinetics ~ Thermodynamics

Consequences of strain as amorphization proceeds

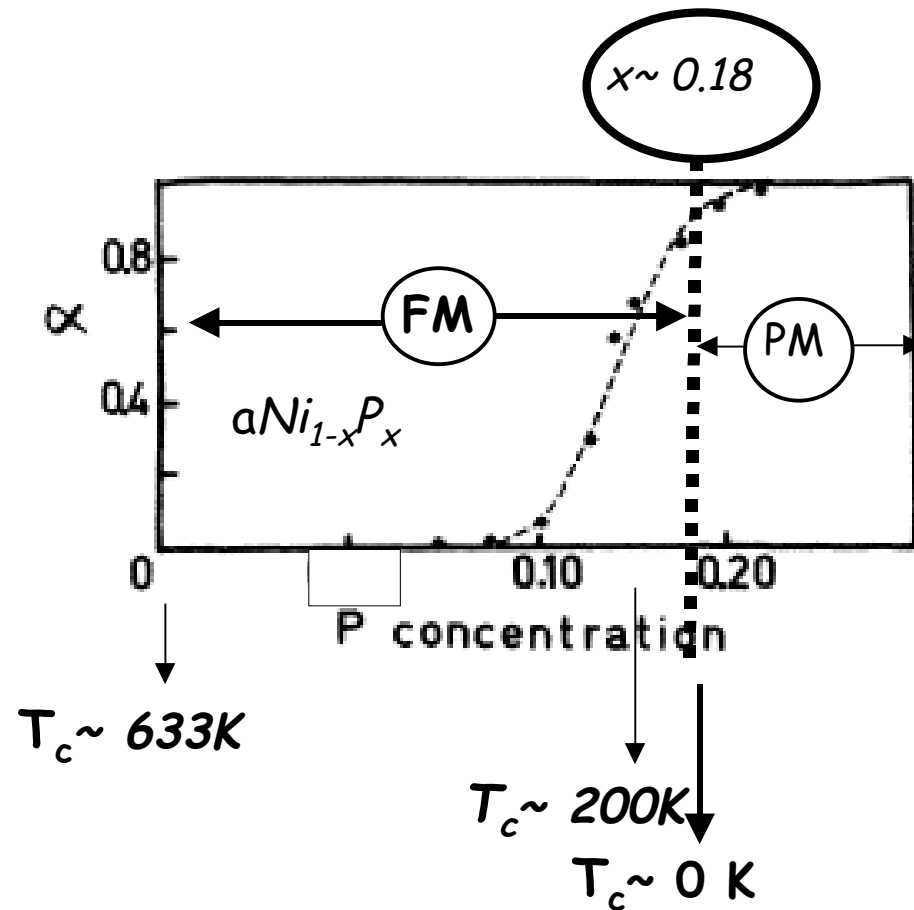
- defect bands :
tearing (1)
- irradi-induced creep & flow:
the chewing-gum effect (2)

MO Ruault, M. Schack et al., 1988

See film



Conduction/magnetic properties: Hall resistivity & EHE in $Ni_{1-x}P_x$:



A. Traverse et al., PRB (1988)

Expected from work on "macroscopic" Metglass:
Magnetism disappears @ $c \rightarrow$ a transformation ?

ture.⁹ The resistivity was measured at 4.2 K after each implantation step and, using ac measurements, the Hall voltage was followed against the magnetic induction at 4.2 K and in some cases at 80 K.

IV. HALL EFFECT IN Ni AND ITS ALLOYS

The Hall resistivity ρ_H is given by

$$\rho_H = V_H t / I, \quad (1)$$

where V_H is the Hall voltage, t the sample thickness, and I the current. An experimental V_H versus B curve, obtained in a magnetic sample such as Ni (shown in Fig. 3 for $x=0.0$) can be described empirically by the following formula in SI units, using the notation of Hurd:¹¹

$$\rho_H = (R_0 B + R_S M) \mu_0, \quad (2)$$

where M is the magnetization in A/m, μ_0 the vacuum magnetic permeability, and B the applied induction in T. The ordinary Hall coefficient R_0 , expressed in $\text{m}^3/\text{A s}$, accounts for the Lorentz force acting on the electrons, while R_S , the spontaneous Hall coefficient expressed in the same units, is a characteristic contribution for magnetic materials. If B is written in terms of the internal magnetic field H_i , Eq. (2) becomes

$$\begin{aligned} B &= (H_i + NM) \mu_0, \\ \rho_H &= (R_0 H_i + R_1 M) \mu_0 \quad \text{with } R_1 = R_0 + R_S, \end{aligned} \quad (3)$$

where R_1 is the extraordinary Hall coefficient and N the demagnetization factor, equal to unity in thin films.¹² For low B values, the slope at the origin of ρ_H versus B is R_1 .¹² When all the domains are aligned, the spin saturation is reached, the magnetization is M_s , and the Hall voltage is linear versus B with a slope R_0 .¹¹

In the low-field regime, the normal Hall coefficient R_0 is equal to $1/n^*e$, in a free-electron model, where n^* is the effective number of charges per unit volume.¹¹ In Ni, R_0 is negative,¹³ indicating electronic conduction.

As shown by Hurd,¹¹ the spontaneous Hall coefficient R_S (negative in Ni) is

$$R_S \propto \rho_0^2 \sigma_H^{(1)}, \quad (4)$$

where $\sigma_H^{(1)}$ only depends on the spin of the system and ρ_0 is its residual resistivity.

Previous Hall-effect measurements on the thin Ni films, performed by Le Bas,⁹ had revealed the existence of an easy magnetization axis inducing a remanent magnetization vector in the absence of applied induction, tilted at an angle to the film plane. This affects both the initial slope of the Hall resistivity (the measured quantities R_1 and R_S become R'_1 and R'_S), and the saturation magnetization (which is now B_s instead of M_s , B_s being the projection of M_s on the applied field direction). These film effects also depend on evaporation conditions and subsequent annealing treatments.⁹

The magnitude of R_0 is affected by alloying at concentrations high enough to modify the band structure and consequently the number of charges per unit volume.¹⁴ Through ρ_0 , R_S is strongly correlated to the disorder in the sample, as experimentally shown in several cases.¹⁴

In the paramagnetic state, which is reached for $x \sim 0.18$ in $\text{Ni}_{1-x}\text{P}_x$,¹ the Hall resistivity must be linear with H :

$$\rho_H = (R_0 + R_1 \chi) H, \quad (5)$$

where χ is the magnetic susceptibility.¹¹

V. RESULTS

The values of R_0 , R'_1 , and R'_S , equal to $R'_1 - R_0$, were found to be negative, in Ni films whatever the P concentration from 0 up to 0.28; hence, in the following we have omitted the negative sign and only presented the absolute values.

A. On pure Ni films

In agreement with the results of Le Bas,⁹ our measured average value of B_s is 0.48 T (the bulk value being 0.6 T). Our values of R'_1 [typically $(10.39 \pm 1.50) \times 10^{-10} \text{ m}^3/\text{A s}$ for $550 \leq t \leq 1950 \text{ \AA}$] agree with those

Effect of microstructure on transport properties (Hall resistivity & EHE)

α = amorphisation rate

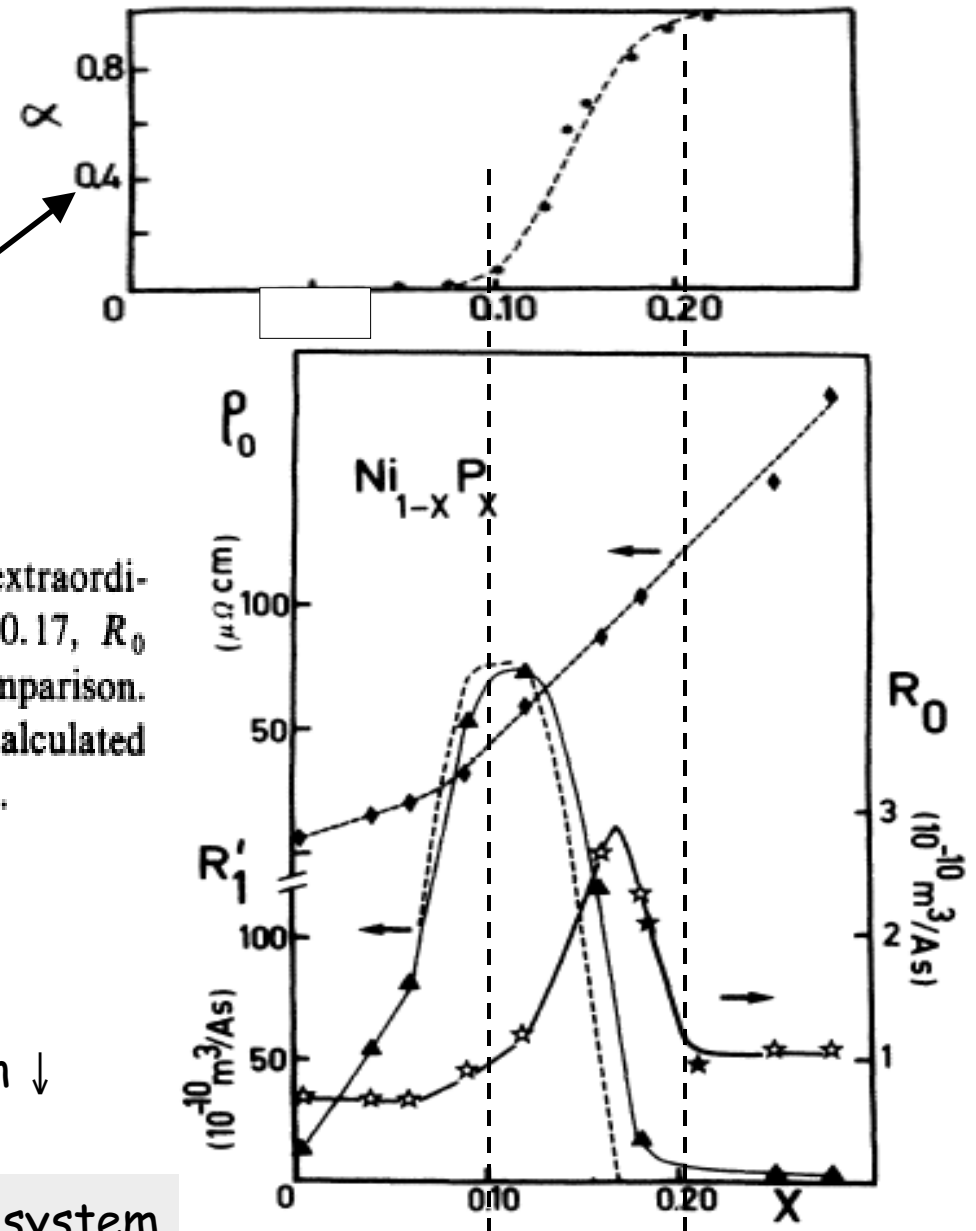
FIG. 4. Residual resistivities ρ_0 , normal (R_0) and extraordinary (R_1') Hall coefficients plotted vs x . For $x > 0.17$, R_0 values of Ref. 21 (solid stars) are also plotted for comparison. Solid lines are to guide the eye. The dotted line is a calculated variation (see text) for R_1' in the range $0.08 < x < 0.17$.

$$R_1' \propto \rho_0^2 \alpha(H)$$

$$\sim (1-\alpha)^2$$

Magnetic contribution ↓
when α ↑

Consistent with inhomogeneous system



*Amorphization via implantation :
the mechanism*

Kinetics vs. Thermodynamics



Ergodicity ? (no random disordering)



Control parameter @ LT =

- atomic displacement rate (+ *solute atom flux*)
- chemical short range order (CSRO)



Varies as $f(T)$

(relate to F.E. curves)

determines stability criterion

Note compatibility with Martin's approach

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(and see refs. in Paolo Ossi's lecture)